MARCH 1955

THE WELDING JOURNAL

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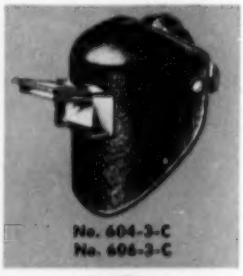

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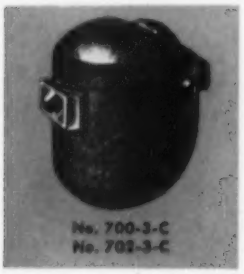

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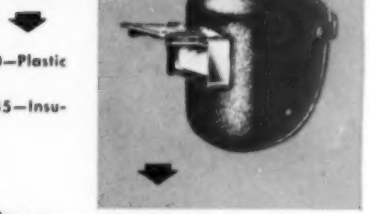

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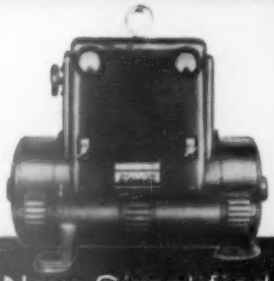

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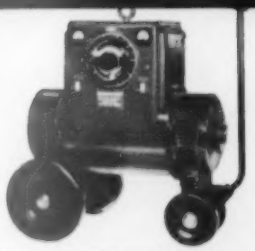

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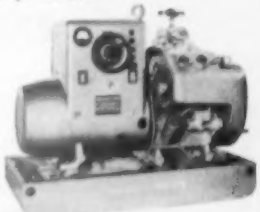

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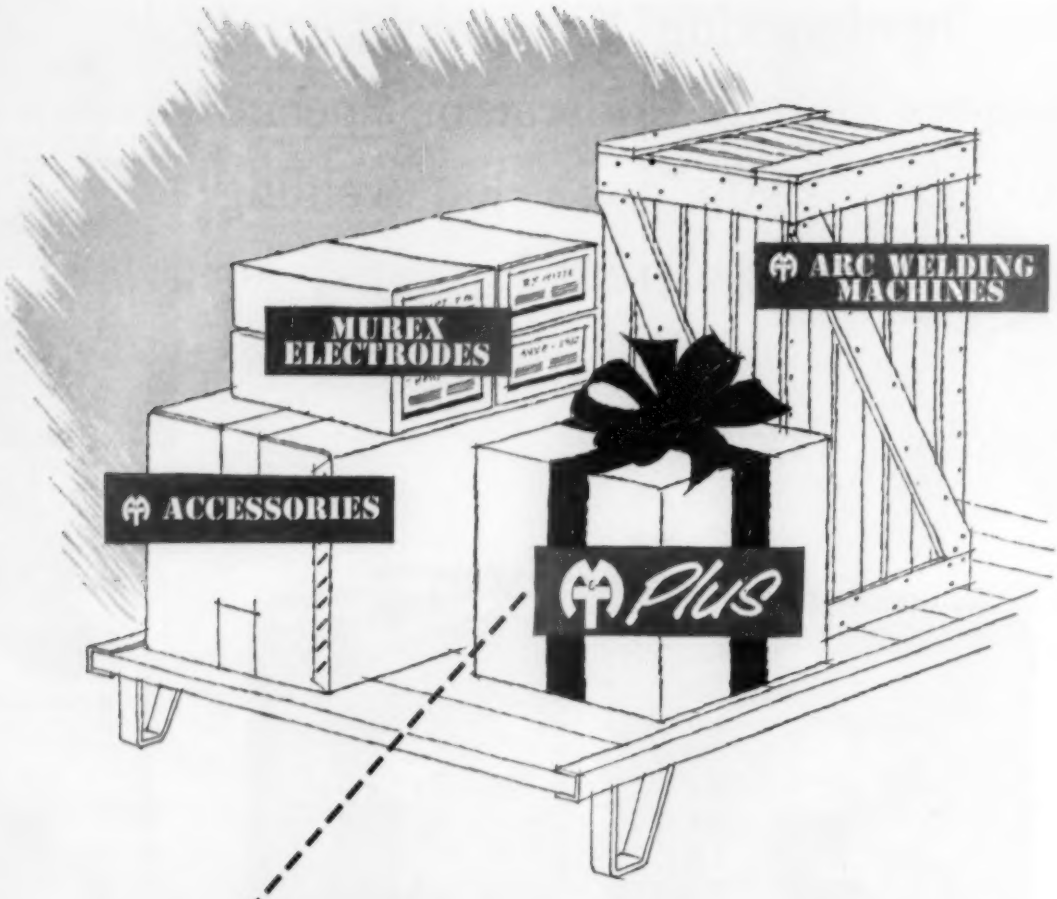

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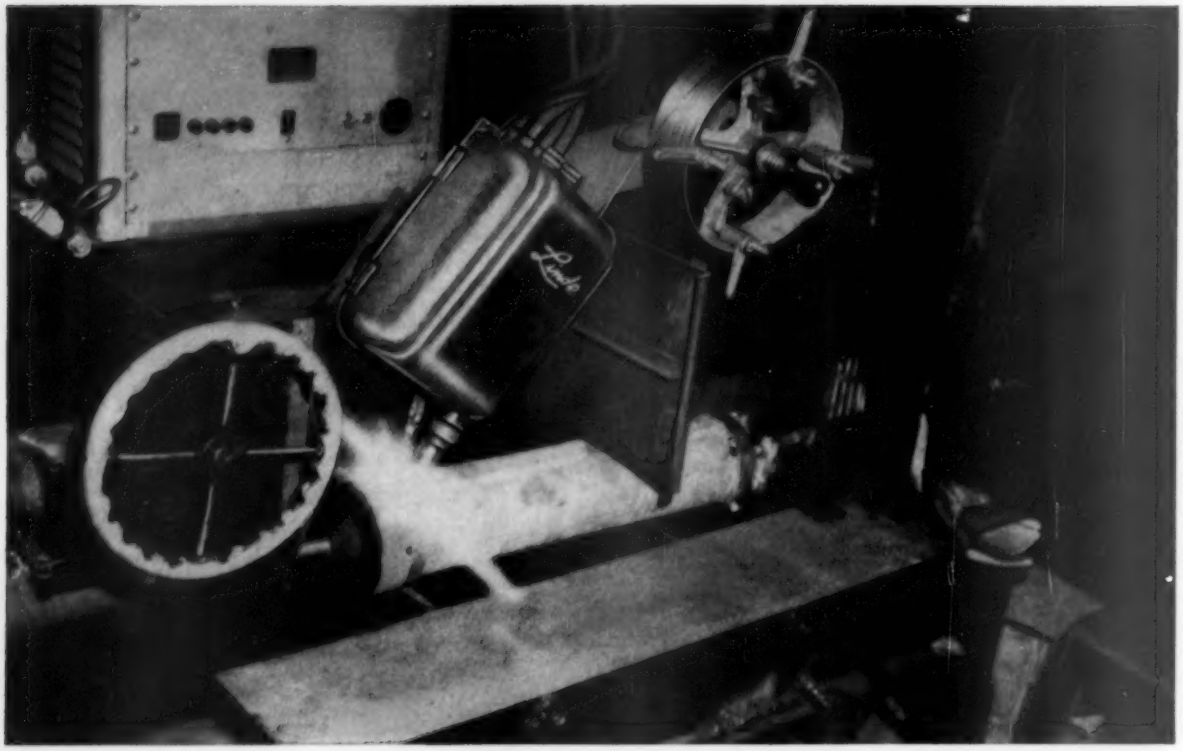

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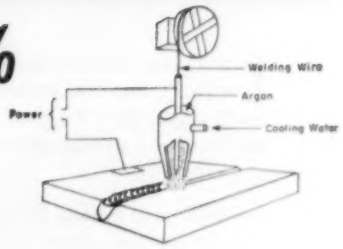

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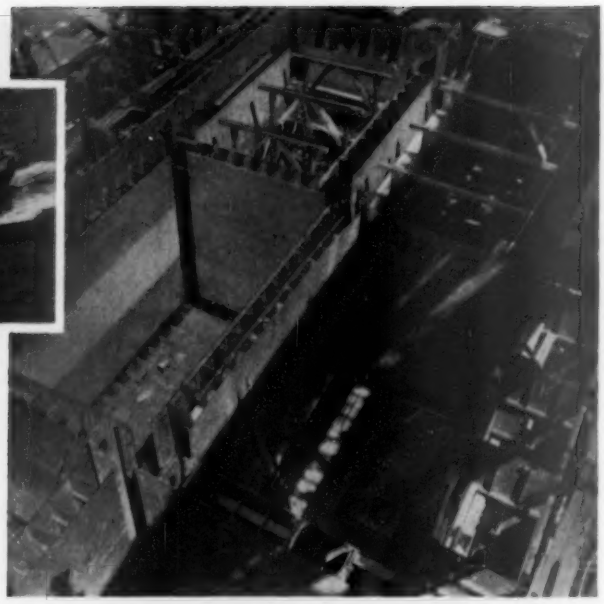

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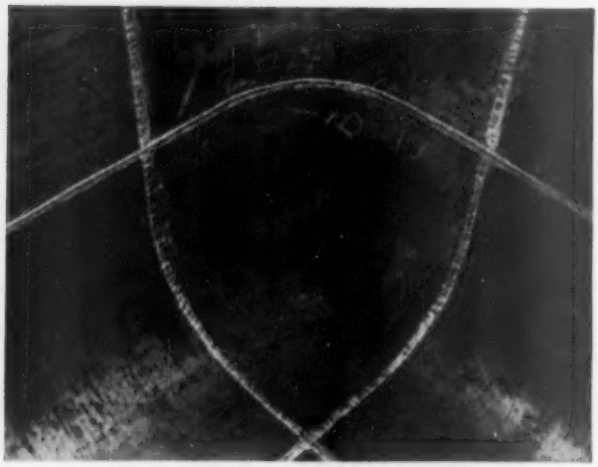

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EDITORIAL

THE VALUE OF RESEARCH

In this day and age, research is accepted by both the lay and engineering public as a must. It is no longer necessary to sell the idea of the effectiveness of research. However, the term itself means different things to different men. Actually, any application of science to the initial solution of a problem may be called research, and properly. The science of welding has come into being by the application to the welding problem of sciences developed in many other fields—sometimes without direct intent. But today welding is itself a science and requires research specifically directed to its problems. This means that the first quick advance sparked by the flash of genius is the exception rather than the rule.

Requirements are manifold, and are becoming ever more stringent, and ever more specific. However, even with the mounting costs incident to the study of any problem, the rewards are still so great that, given intelligent selection and execution, the gamble of being successful is well worth while. This is true whether the research be proprietary leading to the sale of an article or process, or whether it be for the common good. Who is to put a precise figure on the benefit to industry and mankind in general due to the change in allowable design stress for pressure vessels from 9000 psi to 15,000 psi and on to 27,000 psi which can be directly attributed to such research? Who can put a similar value to our present ability to weld stainless steel, aluminum and the less common metals, again the result of research? Of course such precise evaluation is more than difficult, but the order of magnitude of the benefit is so great that the judgment as to the return from the researches is simple, far and beyond any need for such evaluation.

No one can predict the outcome of any research project but the over-all result can be extrapolated from the past. And the indicated path and conclusion is clear: continue research intensively and let it be limited only by the availability of men capable of carrying it out successfully.

Augustus B. Kinzel
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WELDING TYPE 347 STAINLESS STEEL FOR 1100° F TURBINE OPERATION

Special electrode composition and postweld heat treatment employed to produce welds insensitive to sigma phase embrittlement at service temperatures

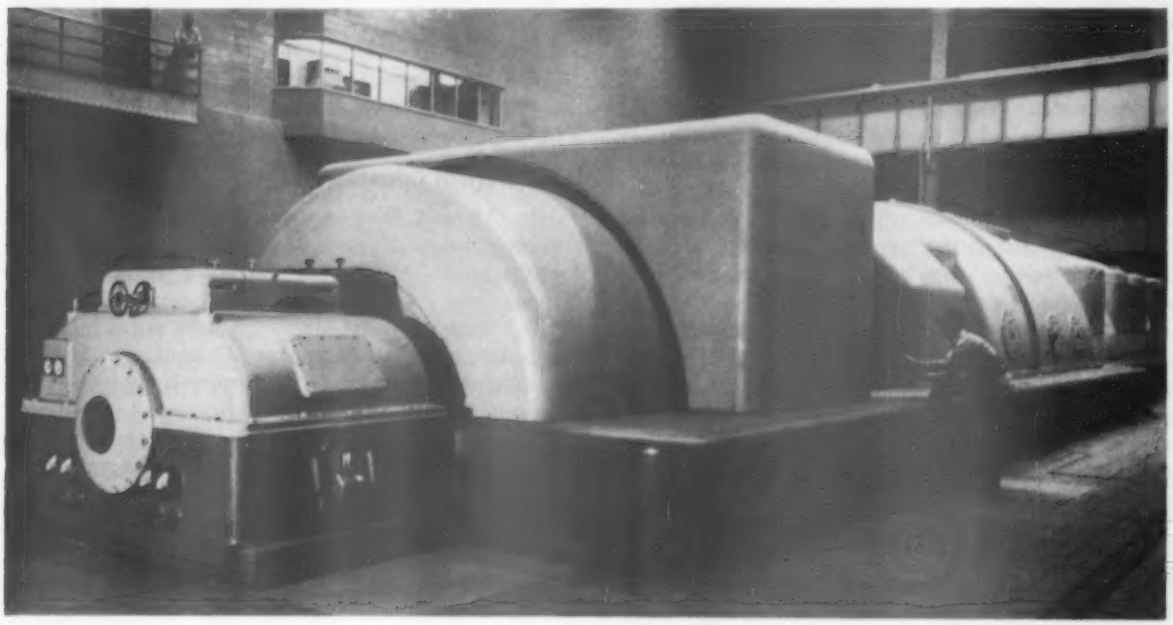
BY R. M. CURRAN AND A. W. RANKIN

Between November 1948 and February 1952, ten large steam turbines totaling approximately 1,000,000 kw capacity were placed in operation at the 1050° F temperature level by the authors' company. All these units utilized austenitic stainless steels of the 316, 321 and 347 families for the major turbine parts such as the inner shells, the stop and control valve casings and heads and the piping between the stop valves and the turbine proper. All the welds in the latter piping were made with Type 347 electrodes with approximately 8 to 10% ferrite content and with no preheat; on the first such unit, most of the piping welds received a 1600° F postweld treatment, but a number of the piping welds on that unit together with all the piping welds on the nine subsequent units received no heat treatment after welding. Although the development of a 1% chromium, 1% molybdenum and ¼% vanadium steel in 1950 resulted in subsequent units for the 1050° F temperature level being constructed of this low-alloy ferritic steel, the operating experience gained on the ten units constructed with the austenitic steels was of invaluable assistance when design work was started on the first 1100° F unit—the Kearny No. 7 unit of the Public Service Electric and Gas Company of New Jersey. On this unit, the step to the higher temperature level resulted in the reintroduction of the austenitic

R. M. Curran is associated with the Materials & Processes Laboratory, and A. W. Rankin is Supervisor, Turbine Structural Engineering, General Electric Co., Schenectady, N. Y.

Presented at the Thirty-Fifth AWS National Fall Meeting held in Chicago, Nov. 1–5, 1954.

Turbine end of the Kearny No. 7 Unit of the Public Service Electric & Gas Company of New Jersey. This is the first large steam turbine-generator operating at 1100° F



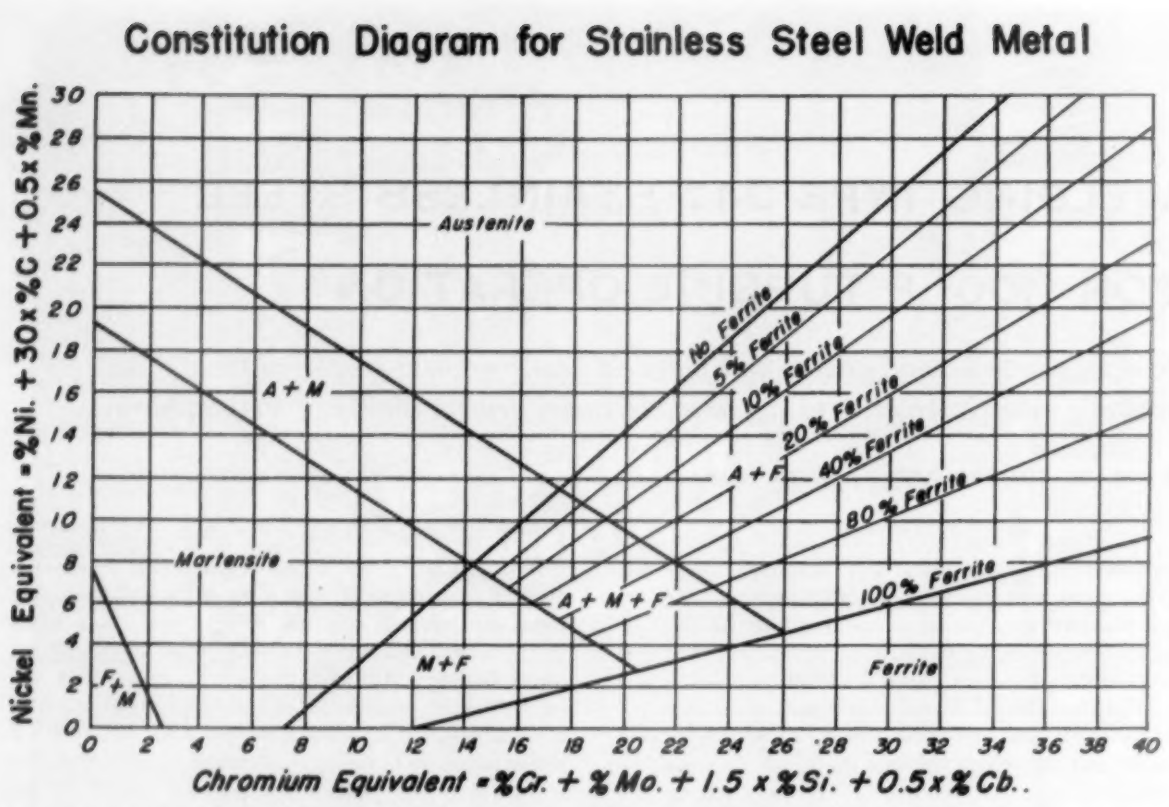


Fig. 1 Constitution diagram for stainless steel weld metal after Anton L. Schaeffler showing the effect of weld metal composition on its structure

stainless steels for the major turbine parts as well as its use for the main station piping.

While essentially satisfactory operation has been obtained with the ten 1050° F units employing the austenitic stainless steels, cracking has occurred in and adjacent to the welds to a relatively greater degree than has been experienced with the ferritic steels. Such cracking has been caused either singly or by a combination of a greater notch sensitivity of the austenitic steels, a slower relaxation of residual welding stresses, or excessive thermal-expansion stresses. In addition, examination of a number of these welds, together with long-time aging tests, have shown that their ductility decreases with time at the operating temperature due to the formation of sigma phase in the deposit itself. While the effects of notch sensitivity residual welding stresses and thermal expansion stresses can be minimized, and perhaps eliminated, by design and erection modifications, the elimination of the sigma formation appeared to be a more difficult problem, and, in particular, appeared to be one whose effects would become more pronounced as operating temperatures reached and exceeded the 1100° F level. Accordingly, a series of technical meetings were held in late 1951, sponsored by the Public Service Electric and Gas Company of New Jersey, to discuss welding electrodes and procedures through which sigma phase formation could be minimized in the austenitic welds to be used in the projected 1100° F units. Investigations being conducted at that time and bearing on this subject by some of the participants in this series of technical meetings have been presented in prior publications.[1–3] The purpose of this report is to describe the investigations conducted by the authors' company in this development of an electrode and welding procedure through which Type 347 welds could be obtained which would be essentially insensitive to sigma phase embrittlement during service at 1100° F and higher temperatures.

Theoretical Considerations

In the early efforts to weld the austenitic stainless steels, considerable weld-bead and crater cracking was encountered. Studies of the effect of weld metal composition disclosed that higher chromium and lower nickel contents in the deposited weld metal produced a greater resistance to this type of cracking. Metallographic studies of such higher chrome–lower nickel weld deposits showed that they contained small percentages of ferrite in the austenitic matrix. The effect of changes in weld-metal composition on the percentage of such ferrite in weld deposits is shown in Fig. 1.[4] With the equivalent chromium and nickel contents of the customary Type 347 weld deposits, an increase in the chromium content or in the content of elements that act like chromium or corresponding decreases in

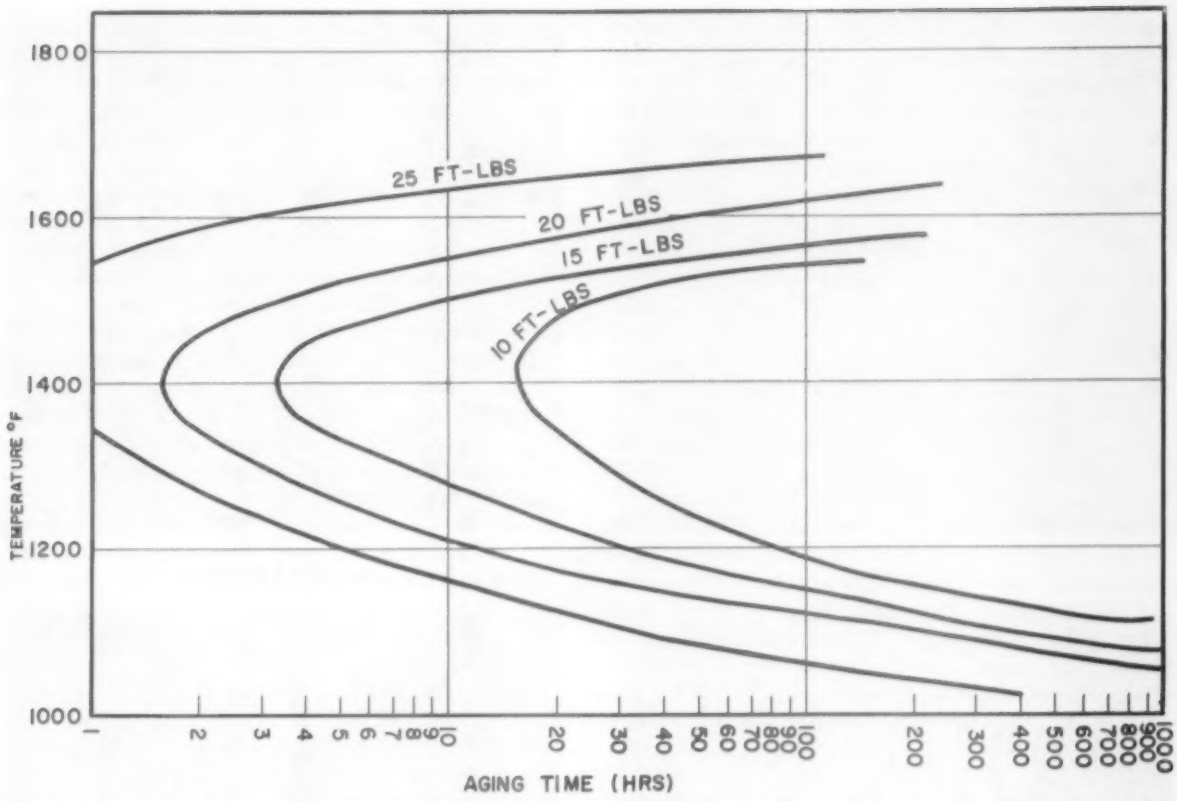


Fig. 2 Effect of aging time and temperature on the room-temperature keyhole-Charpy impact strength of Type 347 weld deposits containing 5–10% ferrite. Initial impact strength 30 ft-lb

the nickel content, tend to increase the percentage of ferrite present in the weld deposit.

While the immunity of such weld deposits to cracking during welding has been accepted for some time, and weld metal compositions of Type 347 have been designed to contain percentages of ferrite, the effectiveness of ferrite in preventing cracking still is not completely understood. The theory has been advanced that the cracking of fully austenitic weld deposits is due to the formation of low-melting silicate films in the columnar grain boundaries,[5,6] and that these films rupture upon shrinkage of the bead immediately after solidification. The effectiveness of the presence of ferrite in preventing this type of cracking has been attributed to the fact that the silicate films probably form around the ferrite particles instead of in the grain boundaries, and hence are made ineffective in so far as cracking is concerned.

Although the ferrite content required to eliminate weld-bead cracking probably is very small, it is necessary to guard against dilution from the fully austenitic base material into the weld deposit making the latter also fully austenitic; accordingly, it has been the general practice to design the weld-metal composition so that some ferrite will be present in spite of base-metal dilution. Inasmuch as the presence of ferrite in the weld deposits did not adversely affect the moderate-temperature properties of the weld metal and had little effect on the corrosion resistance, and since, in addition, the effect of prolonged high-temperature exposure was not understood, few intensive investigations were conducted with regard to depositing sound welds with no ferrite present.

In the postweld heat treatment of Type 347 welds at temperatures in the range from 1200 to 1500° F,[7] however, it was found that the hardness of the weld metal increased and the ductility decreased due to the conversion of the ferrite to sigma phase. Considerable cracking in the weld and in the base metal immediately adjacent to the weld was also encountered during the postweld heat treatment. Since the postweld heat treatment resulted in a decrease in ductility as compared to an increase in ductility for welds in ferritic steels, it was quite obvious that the sole benefit derived from the usual postweld treatment would be relief of the residual stresses induced by the welding operation. In view of the excellent ductility of weldments in Type 347 as-welded, it had been generally agreed that the residual stresses would not be harmful, and that the postweld heat treatment should not be applied in view of the loss in ductility and danger of cracking associated with such a procedure.

Aging tests conducted on weldments of Type 347 at 1100° F and operating experience with steam turbines at 1050° F showed that Type 347 weldments would suffer a loss in ductility in service at these operating

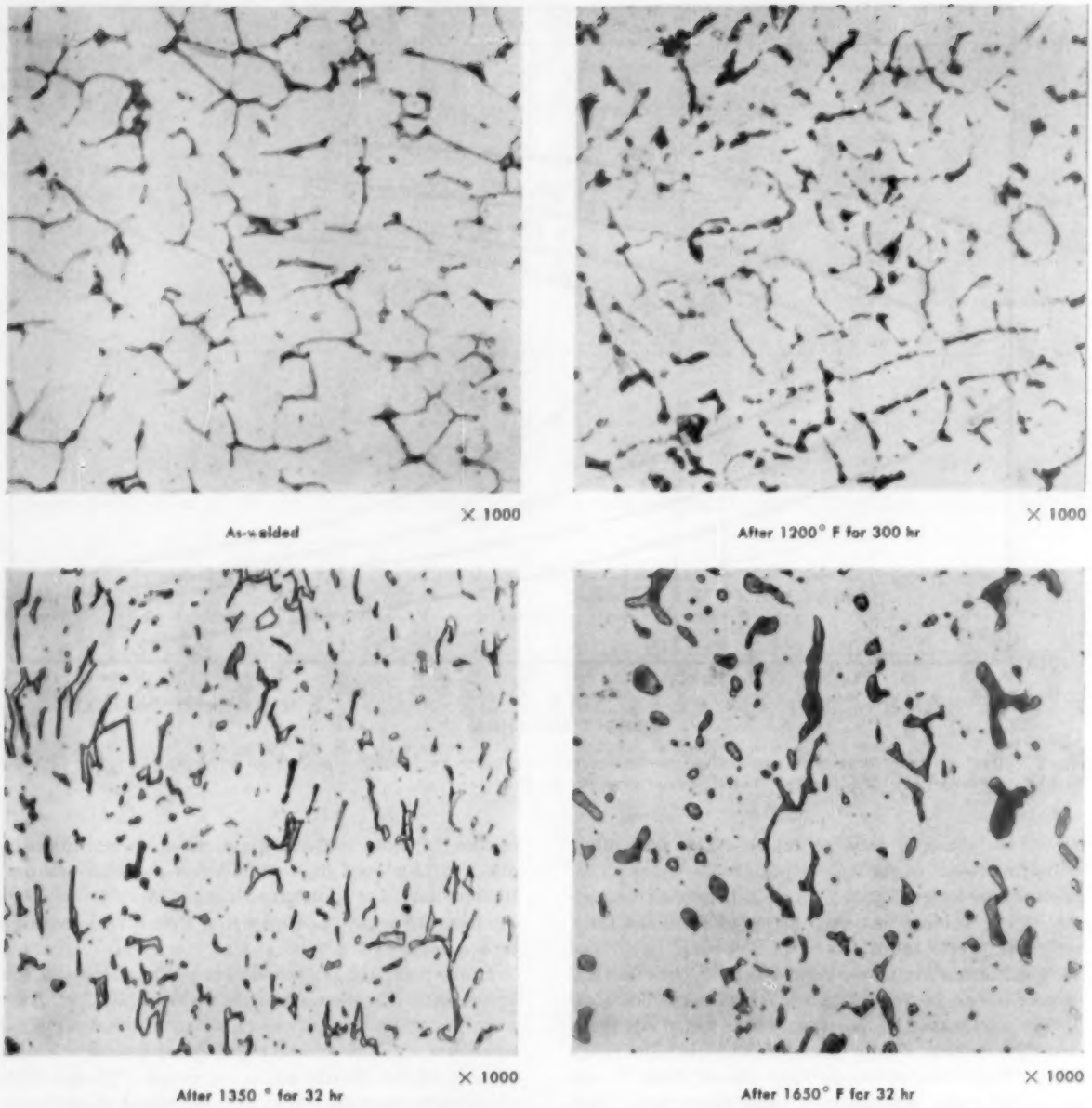


Fig. 3 Photomicrographs showing the structure of Type 347 weld deposits as-welded and after aging at 1200, 1350 and 1650° F. Etched with modified Murikamis reagent. Matrix-austenite, light etching second phase-ferrite, dark etching phase-sigma phase

temperatures. It thus was apparent that elimination of the postweld heat treatment merely postponed the time at which this loss of ductility would occur. An investigation was initiated, therefore, to determine the extent to which the loss in ductility would proceed and, if possible, to determine means of preventing such embrittlement.

Electrode and Heat-Treatment Development

Although material properties such as tensile strength, yield strength, rupture strength, creep strength and the like are directly convertible to allowable design limits, it is difficult in any type of construction to decide upon what basis satisfactory ductility limits can be set. In the absence of suitable criteria to establish such limits, every effort is made to provide as much ductility as possible without undue sacrifice of tensile properties, and an evaluation of the ductility levels achieved is made on the basis of the suitability of similar materials operating under similar circumstances.

Since the Charpy impact test has been shown to be the most sensitive measure of sigma embrittlement, it was decided, in this investigation, to use the room-temperature Charpy impact strength as a measure of the extent of embrittlement of welds at elevated temperature. Although the results of such tests are not

directly convertible to reduction of area or percent elongation, it was assumed, for the initial investigations, that an improvement in the room-temperature Charpy value corresponds to an improvement in the operating ductility of the weldment.

The first step in the investigation was a study of the effect of aging temperature and time on the Charpy impact strength of Type 347 weld deposits. The resulting data are plotted in Fig. 2 showing the time at temperature required for the weld deposit to drop to the indicated level of impact strength from the value of 30 ft-lb obtained in the as-welded condition. This figure is only schematic because of the limited data and because it is peculiar to each particular composition as changes in the ferrite content would undoubtedly shift the figure on both the horizontal and vertical axes. Photomicrographs showing the change in structure of the weld deposit with time at 1200, 1350 and 1600° F are presented in Fig. 3.

The photomicrographs (Fig. 3*a*) of the as-welded structure show a nearly continuous network of ferrite in austenite. At 1200° F (Fig. 3*b*) the ferrite changes to sigma phase without appreciable change in distribution. At 1350° F (Fig. 3*c*) some spheroidization of the ferrite particles accompanies the sigma formation, while at 1650° F (Fig. 3*d*) there is apparently a decrease in the percentage of ferrite present together with spheroidization of the remainder without sigma formation.

It appeared probable from examination of these and other data presented in the literature that the reduction of ductility caused by sigma phase formation is due to the formation of this phase, with essentially no ductility, in an austenitic matrix whose properties are unaffected by the local conversion of ferrite to sigma phase. It seemed reasonable to expect, therefore, that the reduction in impact strength would be a function of the percentage of the fracture path which occurs through sigma phase, and that any steps which would reduce this percentage would also improve ductility. This resolved itself then into the problem of both reducing to a minimum the volume of ferrite and spheroidization of the remainder in order to eliminate the more or less continuous network present in the structures of Figs. 3*a* and 3*b*. Since the 1650° F treatment of Fig. 3*d* was found to both spheroidize the ferrite and reduce its volume, it appeared that by utilizing a low initial ferrite content together with a suitable heat treatment, a minimum of ferrite would be present to convert to sigma, and would be essentially spheroidal in form so that its conversion to sigma would not cause any serious loss in ductility.

In order to test the foregoing hypothesis, a number of Type 347 welds were made with initial ferrite contents ranging from $2^1/_2$ to 12%, with postweld treatments varying from 1800 to 2000° F, and with subsequent aging at 1350° F for 300 hr. The latter aging was intended to simulate the effect of long-time exposure at the lower operating temperatures, and the 1350° F was selected from the data of Fig. 2 which indicated this to be close to the temperature of most rapid sigma conversion. The keyhole Charpy impact strength of all these welds was approximately 30 ft-lb after the postweld treatment but before the 1350° F aging treatment, and the effect of the latter aging treatment on the impact strength is presented in Fig. 4. As a basis to evaluate the efficacy of the lower ferrite contents in conjunction with the high-temperature postweld treatments, a Type 347 weld with an initial ferrite content of approximately 8% and with no postweld treatment would drop in impact strength from an initial value of 30 ft-lb to approximately 7 ft-lb when exposed to this 300-hr aging at 1350° F. The results of Fig. 4 accordingly indicate that by utilizing the lower ferrite contents in conjunction with a postweld treatment in the neighborhood of 1900 to 2000° F, the effects of the 1350° F aging are minimized considerably.

To further substantiate the foregoing results, a number of Type 347 welds were made with deposited ferrite content of approximately $2^1/_2$%, and with a postweld treatment at 1925° F. In order to simulate more closely the aging effects of many years of service at the operating temperature of 1100° F., the welds of this series were aged at 1200° F for periods ranging out to 6400 hr. The results of these tests are shown on Fig. 5, on which are also shown the effect of 1200° F aging on Type 347 welds with between 5 and 10% initial ferrite content aged in the as-welded condition (no postweld treatment). Figure 5 agrees with Fig. 4 in showing that by utilizing a low-ferrite electrode in conjunction with a high-temperature postweld treatment, a marked improvement is obtained in the resistance to service embrittlement.

While the foregoing tests demonstrated a means of minimizing service embrittlement, the use of such a low-ferrite electrode involves the risk of cracking during welding if base metal dilution causes the weld deposit

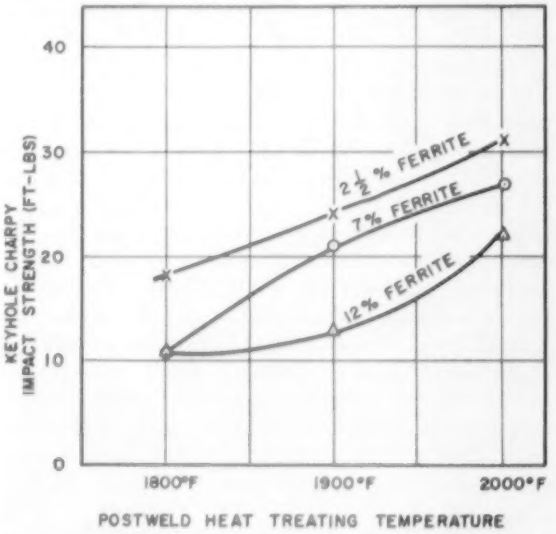


Fig. 4 The effect of initial ferrite content and postweld heat treatment on the impact strength of Type 347 stainless steel weld deposits after exposure at 1350° F for 300 hr

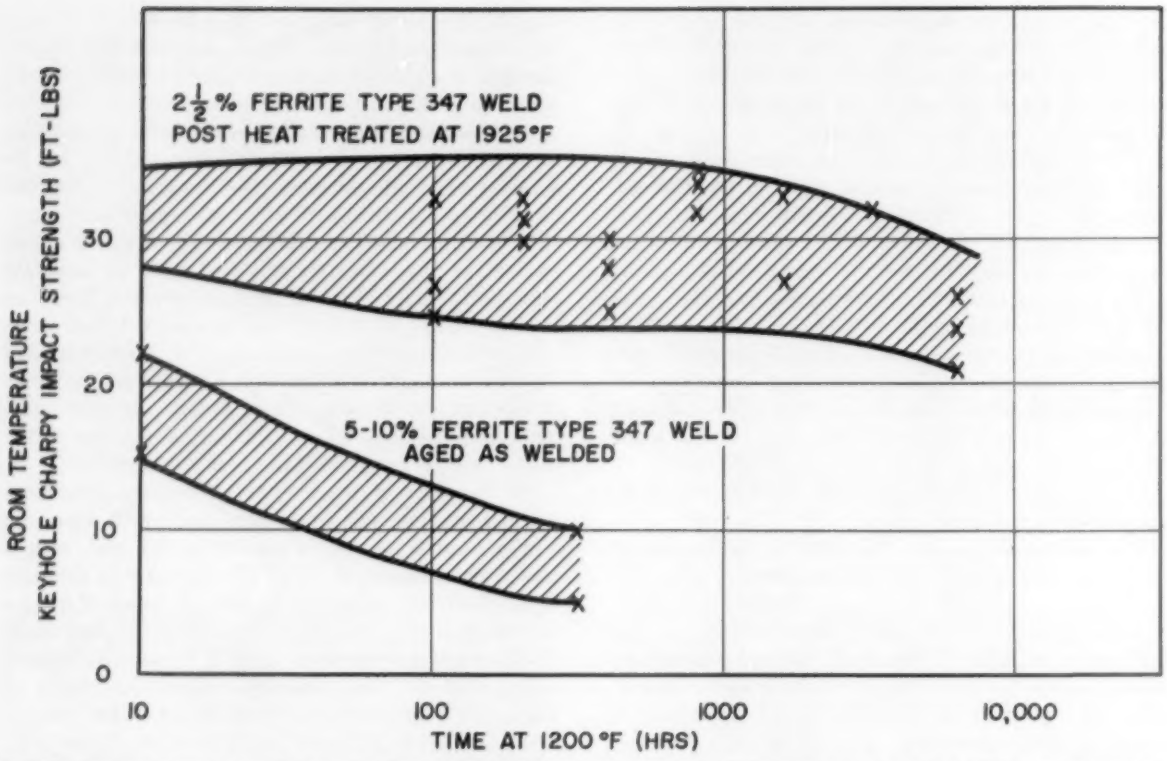


Fig. 5 A comparison between the resistance of postweld heat treated low ferrite and as-welded 5–10% ferrite Type 347 welds to loss in room-temperature impact strength after aging at 1200° F

to become fully austenitic. Previous publications[5,6] have demonstrated, however, that sound welds can be made in the fully austenitic 25% chromium – 20% nickel and 15% chromium – 35% nickel compositions by properly balancing the silicon and carbon contents. Accordingly, tests were conducted on three series of fully austenitic Type 347 welds with varying silicon contents and silicon-to-carbon ratios, and sound welds were obtained on such fully austenitic welds provided the silicon/carbon ratio was 5 or less.

In the 25–20 and 15–35 compositions, and in the foregoing fully austenitic 347 welds, the silicon/carbon ratio was maintained at the desired low level by permitting carbon contents higher than in the standard 347 electrodes. Aging tests on the higher-carbon, fully austenitic 347 welds, however, indicated that the higher carbon content resulted in increased embrittlement tendencies, and that it would be desirable, therefore, to maintain the desired low silicon/carbon ratio by holding to a low silicon level rather than by utilizing a high carbon content. Since it would be impracticable, however, to attempt to maintain a very low silicon content in the relatively large quantities of this electrode which would be needed, it was necessary to compromise somewhat on both silicon and carbon by permitting the silicon to go to 0.50% maximum, and by increasing the carbon maximum to 0.10% from its more usual value of 0.08%. The higher carbon content was considered acceptable for power station steam piping since the atmosphere is not particularly corrosive. The columbium content was specified as 0.60 to 0.90% rather than the more usual columbium specification of ten times the carbon content because the contribution of this element in high-temperature applications is in its improvement of rupture strength rather than in its ability to stabilize carbides to resist corrosion. The final specification for the low-ferrite electrode is given in Table 1.

Procedure Qualification Tests

Qualification tests were conducted on welds made in Type 347 steel using the low-ferrite electrode of Table 1 together with the 1925° F postweld treatment. The results are presented in both Fig. 6 and Table 2; the values obtained in both tensile tests and bend tests are unusually satisfactory, particularly in both tensile and bending ductility.

Table 1—Specification Limits for Special Low-Ferrite, Type 347 Welding Electrodes

Chemical composition, % (deposited weld metal)	
Carbon	0.07–0.10
Manganese	1.50–2.50
Phosphorus, max	0.025
Sulfur, max	0.025
Silicon, max	0.50
Chromium	18.5–20.5
Nickel	9.0–10.5
Columbium	0.60–0.90
Silicon/carbon ratio, max	5/1
Ferrite content*	1–4

* Calculated from chromium-nickel equivalents.

High-Temperature Rupture Strength

High-temperature rupture tests were conducted on these low-ferrite welds with a 1925° F postweld treatment using the Larson-Miller[8] method of trading exposure time for temperature. The results are presented in Fig. 7 on which is also plotted the test data on the high-temperature rupture strength of Type 347 base material.[9] It is evident from this Fig. 7 that the high-temperature rupture strength of the low-ferrite weld deposit is equal to that of the Type 347 base material.

Postweld-Treatment Cracking

While the foregoing investigation indicated that sound welds could be obtained with a low-ferrite electrode with balanced silicon and carbon, and that excellent resistance to service embrittlement could be obtained in such welds when given a 1925° F postweld treatment, the possibility of cracking during this postweld treatment still remained, particularly in the heavy sections to be welded. Quantitative data on such cracking are somewhat limited in scope and applicability because of the difficulty of simulating

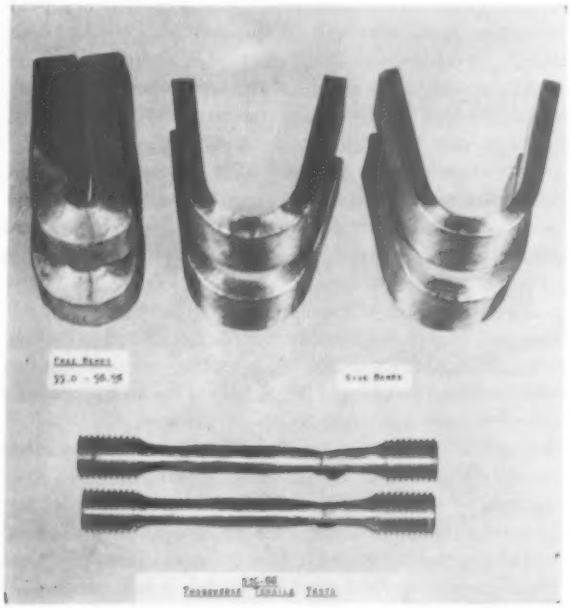


Fig. 6 Procedure qualification tests on restrained Type 347 stainless steel welds made using a controlled low-ferrite electrode and a postweld heat treatment at 1925° F

Table 2—Qualification Test Results

(a)	Transverse tensile tests		
	Ultimate strength, psi	79,000	79,400
	Reduction of area, %	62.6	66.5
	Location of fracture	Base metal	Base metal
(b)	Free-bend tests		
	Elongation, %	55.0	58.5
(c)	Side-bend tests	*	*

* All four specimens satisfactory.

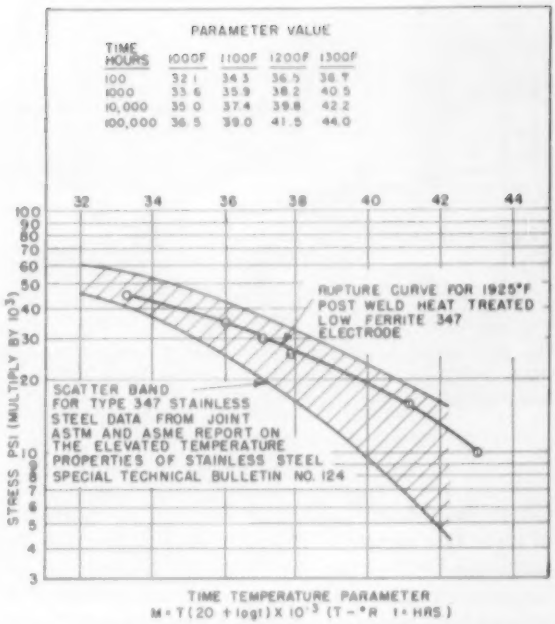

PARAMETER VALUE
TIME HOURS
100 32.1 34.3 36.5 38.7
1000 33.6 35.9 38.2 40.5
10,000 35.0 37.4 39.8 42.2
100,000 36.5 39.0 41.5 44.0
1000F 1100F 1200F 1300F
32 34 36 38 40 42 44
100 90 80 70 60 50 40 30 20 10 9 8 7 6 5 4 3
STRESS PSI (MULTIPLY BY 10³)
RUPTURE CURVE FOR 1925°F POST WELD HEAT TREATED LOW FERRITE 347 ELECTRODE
SCATTER BAND FOR TYPE 347 STAINLESS STEEL DATA FROM JOINT ASTM AND ASME REPORT ON THE ELEVATED TEMPERATURE PROPERTIES OF STAINLESS STEEL SPECIAL TECHNICAL BULLETIN NO. 124
TIME TEMPERATURE PARAMETER
M = T(20 + log t) X 10⁻³ (T = °R t = HRS)

Fig. 7 A comparison of the rupture strength of low-ferrite Type 347 stainless steel weld deposits heat treated at 1925° F after welding with the rupture strength of Type 347 base metal. Time (hr) and temperature (° F) plotted on a parameter basis

weld-joint stresses in a simple test amenable to quantitative analysis. It has been observed, however, that cracking of austenitic welds during the postweld heat treatment is accentuated rather than alleviated by slow rates of heating, and that such cracking is generally intergranular in nature indicating failure at high temperatures. These observations indicate that the cracking is due to the residual welding stresses rather than stresses caused by uneven heating.

The residual stresses present in a highly restrained weld may be expected to approach the level of the room-temperature yield strength, and this level will usually exceed the short-time rupture strength at the stress-relieving temperature. Accordingly, there is a possibility of rupture failure occurring during the stress-relieving treatment, although the realization of this possibility depends on the relative rates of decrease of residual stress and rupture strength during the postweld treatment. While quantitative data on these rates are not available for the complex residual-stress pattern present in welded joints, some data are available from uniaxial relaxation[10] creep and rupture tests, and the data from such tests at 1100° F on Type 347 steel are presented in Fig. 8. This figure indicates that, at 1100° F, uniaxial residual stresses of the initial magnitude used in these tests would eventually result in rupture failure; residual stresses in restrained welds, however, would tend to relax at a slower rate because of the effect of multiaxiality, and accordingly failure would be expected at a somewhat shorter time. At temperatures higher or lower than 1100° F, the point of intersection of the residual stress and rupture strength

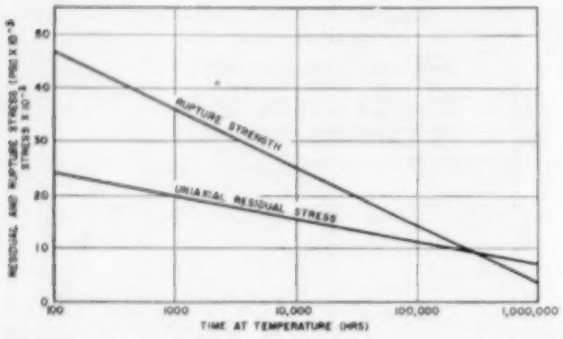


Fig. 8 Curves showing the relaxation of uniaxial residual stress and decrease in rupture strength of Type 347 stainless steel with time at 1100° F

curves would depend, of course, on the rates of decrease of these two factors. Although quantitative data are not available defining this point of intersection for temperatures above 1100° F, the observation that slower heating rates accentuate the cracking of Type 347 weldments during the postweld treatment suggests that the point of intersection of the residual stress and rupture strength curves occurs at a minimum time at some particular temperature. It can be hypothesized, therefore, that a plot of cracking time vs. temperature would be of the general shape shown in Fig. 9, and this in turn implies that the heating above approximately 1200° F should be as rapid as is practicable without introducing excessive stresses due to thermal differences.

On the basis of the foregoing hypothesis, several highly restrained Type 347 weldments were subjected to the following postweld treatment:

1. Heat to 1100° F at 300° F per hour.

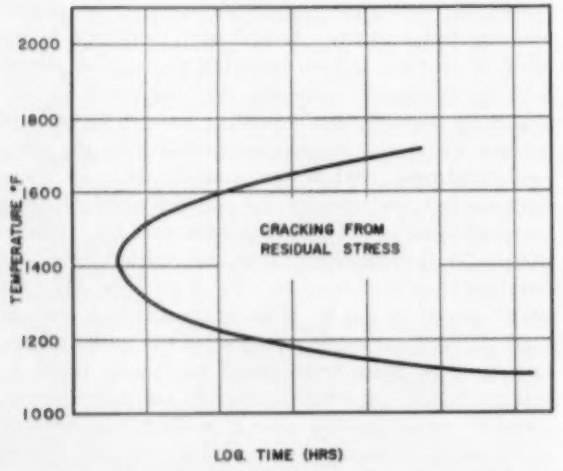


Fig. 9 Hypothetical curve showing time at temperature necessary for residual stresses to cause cracking of Type 347 weldments during postweld heat treatment

2. Hold at 1100° F for 2 hr to permit temperature equalization and relief of the higher residual stresses.
3. Heat from 1100 to 1925° F at 600° F per hour.
4. Hold 2 hr for complete stress relief and ferrite solution and spheroidization.
5. Air cool.

All the test weldments subjected to this heat-treating cycle were found free of cracking upon inspection after the air cool.

Shop and Field Fabrication Experience

The low-ferrite welding electrode and 1925° F postweld treatment described in this paper have already been used for the fabrication of turbine and piping components in several power-generation stations with satisfactory results. Some difficulties have been experienced with both electrode and postweld treatment cracking, but these difficulties have not been of major consequence. The difficulties with the electrode involved some cracking during welding which is not too surprising in view of the low ferrite level which is being maintained. The nature of this cracking indicated that further study is needed concerning the dilution effects of the base metal, and also indicated that detailed attention must be given to the relation between base metal composition and electrode chemistry. The cracking which has occurred during the postweld treatment has also been minor in nature and easily repaired. In particular, these difficulties with both electrode and postweld-treatment cracking have indicated the need for still more intensive investigation of the characteristics of Type 347 welds, but have not been of sufficient magnitude to cast serious doubts on the practicability of the suggested procedure.

An illustration of the type of components on which this welding procedure has been used is given in Fig. 10. This 20-ft-long stop and control-valve assembly, which consists of forged Type 347 valve and pipe components, was welded with the low-ferrite electrode of Table 1 and postweld treated at 1925° F in the Schenectady steel foundry heat-treating furnace. Air cooling was accomplished by withdrawing the car from the furnace. Distortion was held to acceptable values by suitable supporting on the furnace car. Surface cracking occurred on the heavy fillet welds joining the small drain lines to the valve bodies, but no cracking occurred on any of the main structural welds. This assembly was subjected to a second 1925° F postweld treatment following the welding of stubs to the six control valve outlets, and no cracking occurred on any major structural welds during this second furnace heat treatment. A second such assembly was also welded and heat treated in a similar manner with no serious difficulties.

The field welds connecting these valve assemblies to their respective turbines, together with all the welds in the Type 347 main steam and turbine piping, were also welded with this low-ferrite electrode and postweld treated at 1925° F using induction coils with no unusual difficulties.

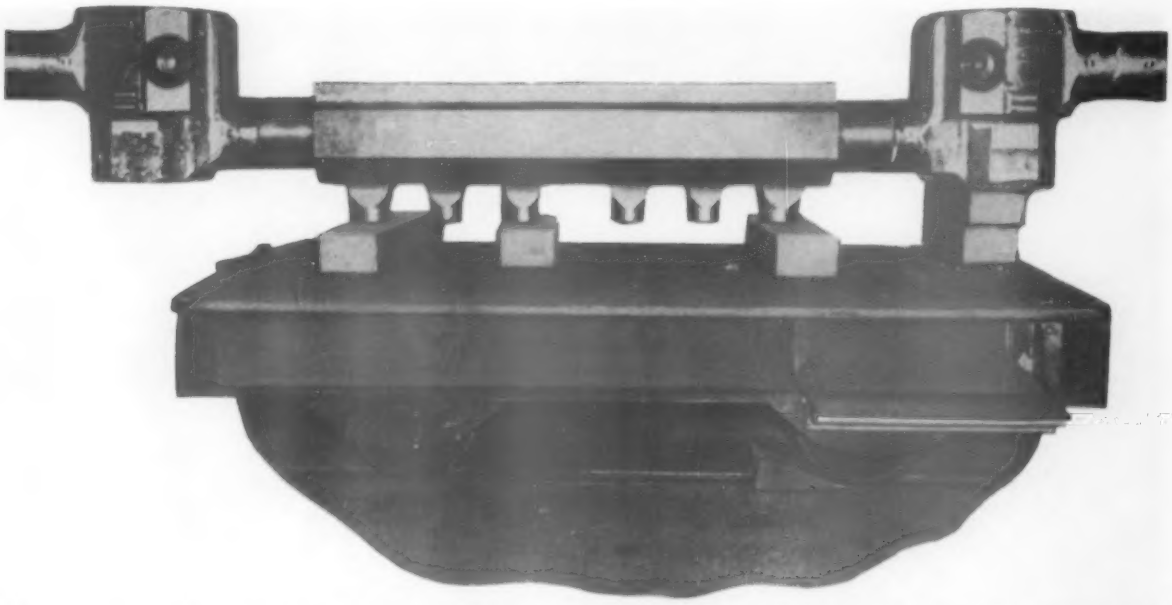


Fig. 10 Photograph showing the stop valve-control valve assembly for the first 1100° F steam turbine. This assembly has an over-all length of 20 ft and weighs approximately 10 tons. Major welds were made between the stop valves and control valves

Conclusions

On the basis of the results presented in this paper, the following conclusions may be drawn:

1. Sound welds can be produced in Type 347 stainless steel using an electrode with a ferrite content in the range of 1 to 4% and with the chemical composition balanced in accordance with Table 1.

2. The use of a 1925° F postweld treatment on welds made with such electrodes results in an appreciable increase in their resistance to embrittlement caused by long-time exposure to elevated temperature.

3. Highly restrained austenitic weldments can be heat treated successfully in the range of 1900 to 2000° F provided a rapid heating rate is used.

4. The long-time, high-temperature rupture strength of low-ferrite welds heat treated at 1925° F is comparable with the corresponding long-time, high-temperature strength of Type 347 base material and Type 347 welds made with the usual commercial compositions.

This paper would not be complete, however, without emphasizing that the authors look on this investigation as only another step in the development of an electrode and welding procedure from which sound reliable welds can be obtained in the austenitic stainless steels. Specifically, this step was intended only to cope with the problem of sigma embrittlement, and even in this isolated area the proof of success must await the results of years of service at the 1100° F and higher temperature levels. In addition, fundamental studies are needed to explain more quantitatively the causes of cracking during the postweld treatment. That which has occurred during the shop and field applications of the electrode and postweld treatment described in this paper has been acceptable on the basis of its being a low price to pay for elimination of sigma formation, but production schedules demand that such cracking be further minimized. In addition, further studies are needed to more adequately explain the service cracking which has occurred in some of the austenitic steel weldments operating at 1050° F as discussed in the Introduction. While it has not been possible to ascribe these to any single cause, there is considerable reason to believe that the complete relief of residual welding stresses afforded by the high-temperature postweld treatment together with the minimizing of the sigma embrittlement will appreciably reduce the incidence of this cracking.

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Spraying of shaft with nickel-chrome-boron (60 Rc) alloy

PRACTICAL HARD FACING WITH FUSED SELF-FLUXING METALLIZED COATINGS

Metallizing technique described by author involves four different stages: preparation of surface, spraying of deposit, fusing of deposit and finishing

BY HARVEY S. MILLER

The process of hard facing by metal spraying is commonly known as "Spraywelding," as it is a combination of spraying and welding. This process will be referred to by that name in the future.

Four stages of technique are involved and will be taken up, respectively, in the following order: Preparation of the Surface; Spraying of the Deposit; Fusing of the Deposit; and Finishing.

I. Preparation of the Surface

The surface preparation for spraywelding is even more important than for regular metallizing, and cleanliness is a must. The part to be spraywelded should be prepared as follows:

(*a*) The part should be undercut by the amount of the overlay required. A minimum overlay of 0.015 in. per side should be used, although we have applied as little as 0.008 in.; and a maximum of 0.090 in. should be used, although we have applied as much as 0.175 in. on a side on a round shaft for special applications. The overlay should be at least 0.010 in. thicker per side than the maximum allowable tolerance before the part is considered to be worn too much for further use. All sharp edges must be rounded off. In the case of flat plates that are to be spraywelded to the edges of the plate, a file should be drawn over the edges and the preparation should go around the corners for at least $^1/_8$ in. In the case of shafts, the corners on the undercut section should be rounded with at least $^1/_{32}$ in. radius. We find that if the corner is machined at 60 deg with the vertical, then the radius is not necessary. The undercut on shafts should go at least $^1/_4$ in. beyond the required hard faced area when practical.

(*b*) All parts should be cleaned off with a solvent or degreased before gritblasting. If a worn shaft is being rebuilt and shows up harder than 32 Rc, then it should

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be annealed before gritblasting. Any part that does not show a frosted appearance after gritblasting should be annealed and regritblasted. Any previously applied facing should be removed entirely before preparing for redepositing the new facing. Under some circumstances parts previously hard faced may be patched up without removing all of the facing, but in general the above holds true.

(*c*) The undercut section should then be gritblasted with either a chilled iron angular grit or an aluminum oxide grit (never use sand, as it leaves a powdery deposit). The author finds that a 50-50 mixture of 25 mesh and 50 mesh grit makes a good combination. The 25 mesh grit raises the surface enough for a good grip and the 50 mesh roughens up the smooth sides between the valleys and crests. The gritblasting is a very important part of the whole operation and should be meticulously performed, especially in the corners of the undercut. The area not to be hard faced should be masked off before gritblasting with a good blasting tape. The gritblasting on shafts should go at least $^1/_8$ in. beyond the undercut section to prevent peeling back or lifting during fusing.

(*d*) Where gritblasting facilities are not available, fine threading may be used. About 80 threads per inch and not over 0.010 in. deep. With this method, however, there is too much danger of not getting the metal clean in between the threads; also if the threads are too deep, air pockets will cause porosity.

Spraying of the Deposit

There are two principal methods of spraying the self-fluxing alloys, namely by means of (*a*) a wire gun or (*b*) a powder gun.

(*a*) The wire gun is commonly known as a metallizing gun of which there are several manufacturers both in the United States and abroad. The wire is fed into the flame either by means of a high-speed air turbine geared down or by means of a small horsepower constant speed electrical motor geared down. The nickle-chrome-boron alloys are too hard and brittle to be drawn into wire form, and therefore cannot be used in a metallizing gun excepting in the form of cast wires centerless ground in 3-ft lengths. This is expensive and impractical. However, a method of mixing the powdered alloy with a highly volatile plastic, which leaves no solid residue after burning, and then extruding this mixture into $^1/_8$ and $^3/_{16}$ in. wire sizes was developed. Both major United States manufacturers of metallizing guns are now producing these wires and they are available to all owners of metallizing guns which feed either one or both sizes. The author does not believe that this wire is being made in 11 gage sizes as the wire does not have too much tensile strength and is fairly easily broken if the operator is not careful. There is approximately 20% plastic binder in each pound of wire. Standard metallizing gas pressures are used.

(*b*) There are several different types of powder guns but the principle is approximately the same for all. The powder is fed from a hopper or container by means of a carburetor which is nothing more than a venturi. This allows an air jet to pick up the powder and propel it out through a hose to a gun where it passes by a flame, is heated to its plastic range and then blown onto the prepared surface. The most economical speed of spraying is 3 lb per hour at 80% efficiency. Faster deposit may be made; however, the spraying efficiency goes down.

On flat surfaces the gun is held in the hand and passed back and forth over the surface with a slight oscillating motion with the tip approximately 6 in. away from the

Fig. 1 Model C Spraywelder set up on pedestal type stand

Fig. 2 Powder torch used for spraying and fusing at same time being demonstrated on small piece of angle iron. Note powder control valve under powder hopper. The two control valves on right are for fuel gases. They pick up the powder on the way through to the tip where the powder is melted and deposited, and fused to the base metal at one time

work as in metallizing. The first 0.003- to 0.005-in. deposit should be made rapidly so as not to oxidize the base metal. Standard metallizing procedures can be used from then on, being careful not to put on too heavy a layer or to overheat the surface, as the overlay will lift. Spray over the edges of the plate on the previously prepared surfaces.

Fig. 3 Steps in spraywelding from left to right: undercut and steel grit blasted; sprayed; fused; finish ground

On shaft work the gun is usually set up on the carriage of a lathe and the work set up to rotate in the chuck of the lathe. The work is rotated and the gun travels back and forth automatically with the tip 5 to 6 in. away from the surface. The first passes should be applied by hand and should be made rapidly, moving back and forth until the area is completely coated with at least 0.003 in., then normal speeds can be resumed, depositing approximately 0.010 per pass. The author prefers light fast passes for heat equalization.

When spraying stainless steel the work should be preheated to about 400° F before spraying begins, which will compensate for the greater coefficient of expansion of the stainless steel.

Preheat is also necessary when spraying on internals on any metal. External heat should also be maintained while the internal part is being sprayed, otherwise as the outside surface heats up due to multiple passes on the inside, it will expand away from the first layer on the inside. This will cause pull-back during fusing. The preheat on carbon steels should be from 200 to 300° F; if higher preheats are used before the first pass is made, oxidation will occur under the overlay and this will also cause pull-back during fusing.

There is a new type of powder gun now available which was developed in England and is being distributed by a well-known United States manufacturer of hard facing materials. This gun sprays and fuses the powder at the same time. It is practicable for thin and heavy coatings on flat surfaces and valve seats and works extremely well where high production

Fig. 4 Fusing shaft with the acetylene torch after spraying with nickel-chrome-boron (60 Rc) alloy

is involved. In the author's opinion, it is not too desirable for round shafts.

Fusing the Deposit

There are three methods of fusing the nickel-chrome-boron alloys, namely (*a*) induction heating, (*b*) controlled atmosphere furnace and (*c*) torch or open flame.

(*a*) *Induction heating* has only been used on high production of small parts and is mainly in the experimental stages. The author has not had a chance to look too deeply into this type of fusing; however, he believes that it has great possibilities on large diameter shafts and very small parts. We have fused small sawteeth by this method that have been first sprayed and they worked out very well.

(*b*) *Controlled Atmosphere Furnace.* We have done quite a bit of this type of fusing. The atmosphere can either be dry city gas or dry hydrogen, preferably hydrogen. All of our work has been done with city gas as we have no hydrogen furnace available. The parts are put into the furnace at approximately 1800° F. The furnace is then brought up to 1900° F in the case of the No. 6 alloy, 1950° F in the case of No. 5 alloy; and 2025° F in the case of No. 4 alloy. The No. 6 alloy is approximately 70% nickel, 16% chrome, 4% boron and 60 Rc; the No. 5 alloy is approximately 76% nickel, 13% chrome, 3% boron and 50 Rc; the No. 4 alloy is approximately 80% nickel, 11% chrome, 2½% boron and 40 Rc. The parts are held at their respective temperatures for approximately three minutes. The temperatures mentioned above are at the beginning of their plastic ranges and therefore will not actually flow the material. Any excessive vibrations while the alloys are at these temperatures will cause what looks like marine borer tracks as the alloy will try to form a ball. The furnace is then dropped to 1700° F after which the parts should be pushed into the slow cooling section of the furnace.

The 400 series steels, "K" Monel and Inconel "X" require slightly different procedures. These are alloys which have become standard in pumps and valves due to their good wear and corrosion resistance. Using ordinary methods of hard facing, these materials present

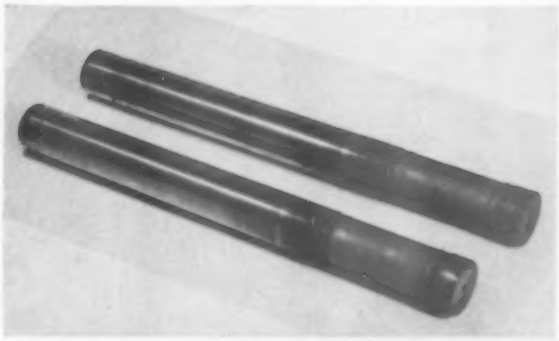


Fig. 5 Duriron pump shafts spraywelded with nickel-chrome-boron (60 Rc) alloy. Note smooth finish before grinding. After finish grinding, these shafts were then machined to correct dimensions. This is a production type of sequence of operations

Fig. 6 Machine spindle spraywelded with nickel-chrome-boron (60 Rc) alloy and torch fused in lathe with acetylene torch; spraywelded areas were finish ground to 0.0001 in.

many problems due to checks and cracks. Using the spraywelding techniques and furnace fusing in a controlled atmosphere dry hydrogen furnace, no difficulties are encountered. The procedure for fusing is the same as was used before except that instead of dropping the temperature to 1700° F immediately after fusing and then taking the part out, you drop the temperature 100° F the first hour and 50° F per hour until 1600° F is reached; then the parts are slow cooled in the cooling section. The results are a homogeneous coating with no checks or cracks.

(*c*) *Torch or Open Flame.* This is the most common method of fusing the self-fluxing alloys on shafts. Oxy-acetylene multiflame torches are used; however, oxy-propane torches seem to give better results on the larger diameter shafts as the flame is softer and envelops more of the shaft with burning gases, thus producing its own reducing atmosphere.

Fusing by torch should take place immediately after spraying for two reasons.

1. The parts have been partially heated by the spraying and the operator does not have to be as careful in his postheat for removing the chill before fusing and he has also saved time and fuel.

2. If the parts lay around unprotected, a thin film of rust could develop and creep under the overlay, thus causing lifting during fusing and porosity by absorbed moisture.

The parts should be brought up slowly all over by manipulating the torch quite rapidly over the entire surface. Shafts or symmetrically round parts should be

Fig. 7 Spraywelding Diesel valves with nickel-chrome-boron (50 Rc) alloy

Fig. 8 (*Left and center*) Worn Diesel valve and valve spraywelded with nickel-chrome-boron (50 Rc) alloy. (Both were in service 5000 hr before pictures were taken. Spraywelded valve showed no appreciable wear.) (*Right*) Valve spraywelded with same alloy and finish machined

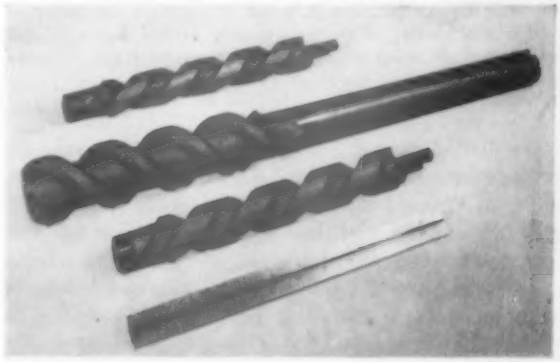


Fig. 9 Imo pump rotor furnace fused. Parts were sprayed while rotating in centers. Nickel-chrome-boron (60 Rc) alloy was used. These parts were furnace fused due to hard facing. No finish is required

Fig. 10 Here is a copper thermocouple well which has been sprayed with nickel-chrome-boron (60 Rc). The overlay was bonded in a controlled atmosphere furnace at just over 1900° F

Fig. 11 Glass plungers spraywelded with No. 4 alloy and torch fused. Very high polish obtained by machining with No. 883 carbide tool

Fig. 12 Oil well valve seat being sprayed on seat area. Valve weighs 150 lb, so had to be preheated before spraying and during spraying in order to equalize temperature and prevent lifting of sprayed metal. Nickel-chrome-boron (50 Rc) alloy being sprayed through powder gun. Valve made of 18–8 stainless steel

rotated while fusing. Large parts do not have to be brought up to a red heat all over but should have the adjacent areas at a red heat while fusing. Whenever possible, fuse the edges before proceeding with the final fusing operation.

Large sections should be entirely surrounded by a refractory material leaving only enough opening to manipulate the torch. After fusing, the opening should be closed and the part allowed to slow cool.

Small sections should be slow cooled by wrapping them in asbestos or dropping them into a bucket of powdered asbestos or lime.

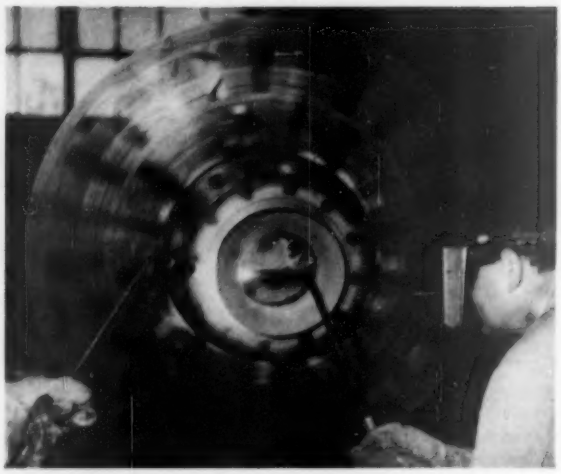


Fig. 13 Torch fusing with acetylene while equalizing heat on outside area with large propane tip. Valve rotating at approximately 50 rpm. Outside dimensions 2 ft in diameter. Seat diameter 1 ft ID

Make sure that you have enough heat available for the job. It is very embarrassing to lose your heat in the middle of fusing and run for more tanks; or to get your part up to just below the fusing temperature and find that you need a larger tip or that you need someone with another torch to help bring up the temperature for fusing the alloy.

We have a "tell-tale clue" in fusing this alloy. The tell-tale clue is the reflection of the fusing torch on the fused surface. When you see this reflection you should move the torch along so that one-half of the reflection of the end of the tip is always invisible. In this way you will not overheat the alloy and thus cause it to flow. The author will outline the fusing of a shaft which has been sprayed in an area which is not on an end.

1. Heat the entire area to about 800° F while rotating.
2. Heat one end of the shaft about 1 in. away from the end of the overlay to a bright red.
3. Move along the shaft to the overlay, keeping the bright red heat.
4. Hesitate long enough on the end of the overlay until you see the alloy taper out to a smooth curve; at this point you will see the tip reflection.
5. Move to the other end of the shaft, using the same procedure.
6. Pass the torch over the complete overlay several times during the above procedures.
7. After the second end has been fused continue along the shaft until the entire shaft has been fused.
8. Play the torch along the area of overlay until an even heat is observed along the shaft.
9. Slow cool by means of asbestos, lime or built up furnace. On very long areas and short areas, of course, the above procedures will have to be modified slightly.

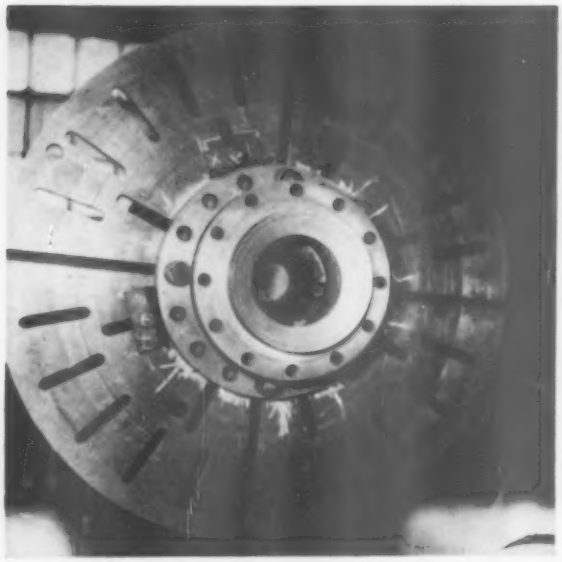


Fig. 14 Seat completely fused and just before removing from face plate for slow cooling by wrapping in asbestos. Very little machining was required. Overlay was kept within 0.030 in. of finish size

Fig. 15 Composite layers of nickel-chrome-boron (60 Rc) alloy and 18–8 stainless steel being sprayed on screw conveyor from large plastics injection molding machine. The nickel-chrome-boron alloy was sprayed with powder gun; 18–8 stainless steel was sprayed with wire gun at same time. Torch fused. The area shown had to be undercut 0.350 in. on diameter

Finishing

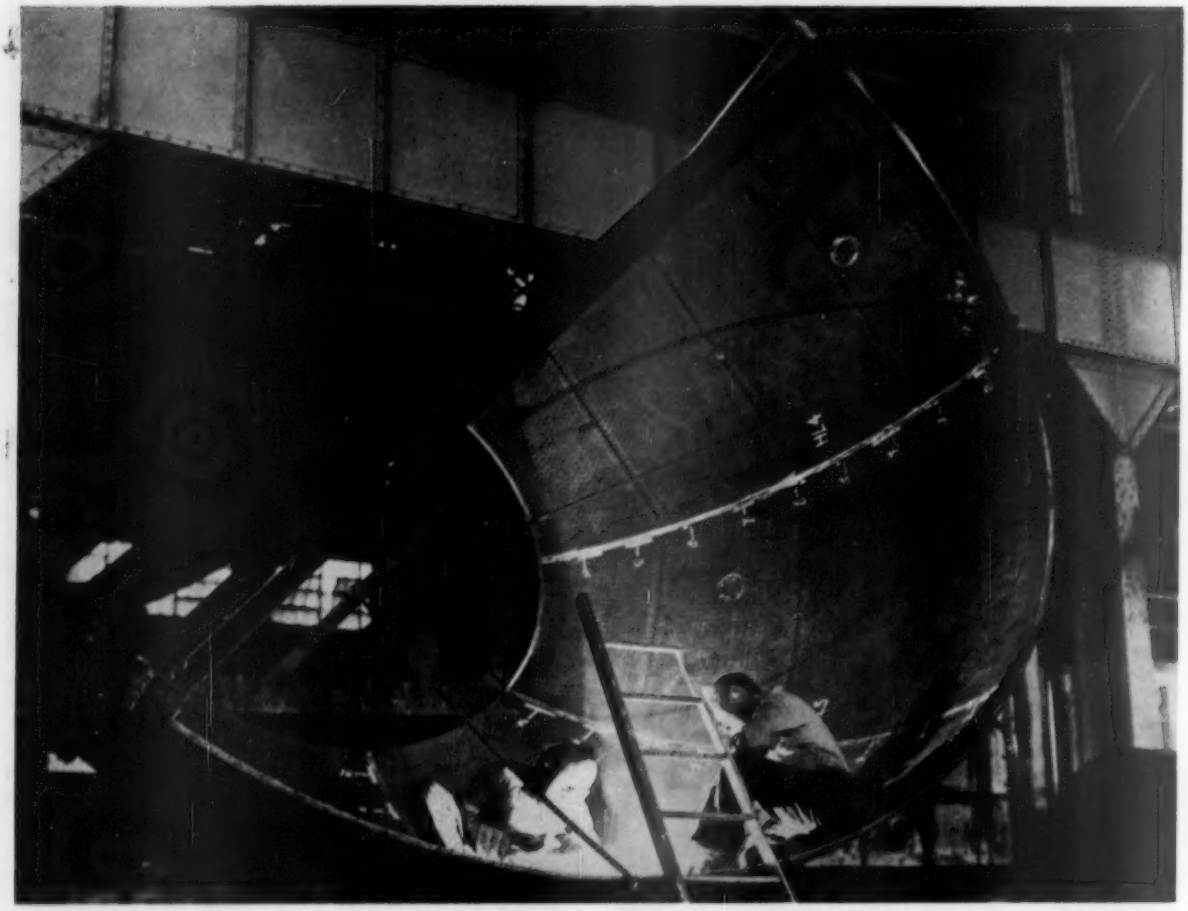
(*a*) *Machining.* Alloys No. 4, 5 and 6 are very easily machined if the correct procedures are used. In the case of the No. 6 (60Rc) alloy, finish grinding is recommended, after rough machining to within 0.015 in. of finish size. No. 883 Carboloy or equal, ground with slight lead and rake angles, should be used. Fine feeds and slow surface speeds are necessary; the tool should be held approximately 2 deg below center. For roughing, fairly heavy chips can be taken, i.e., 0.030 in. at approximately 25 sfm. If you are machining a hardened area in between a soft area, proceed from the soft to the hard, otherwise the tool when it leaves the hard area will dig in due to pressure. Light, fast cuts should be used for finishing. This will give a sort of burnishing effect as the cutting edge of the tool will round over. Remove approximately 0.001 in. per pass when finishing.

Fig. 16 Spindle from automatic screw machine hard faced in wear section with nickel-chrome-boron alloy (50 Rc). Part was grit blasted and set up on turntable. This was sprayed with powder gun by hand and torch fused with propane. Part was then finish ground

Fig. 17 Hard facing of Type 316 stainless steel chemical agitator shaft. Note fire-brick furnace constructed on the lathe. Two large propane torches were used for fusing, with third man located at tail stock end of lathe to move center in and out as shaft expanded during fusing and contracted during cooling. Shaft is $5^1/_2$ in. (solid) OD where hard faced and 30 in. long. Over-all length is 16 ft

Fig. 18 Stainless steel thermo-well 3 ft long by $^3/_4$ in. diam being sprayed with alloy No. 6 (in plastic wire) through an electric-type metallizing gun. Thermo-well is set up in an electric turntable; metallizing gun is mounted on an automatic straight line cutter with tracks running parallel to thermo-well

(*b*) *Grinding.* In general, green silicon carbide wheels with H, I or J hardness, and 24 to 60 grit for roughing and 80 or finer for finishing should be used. Light, fast cuts approximately 0.001 per pass should be taken. All grinding should be done wet if possible. Most of the grinding wheel manufacturers make similar wheels; however, we happened to have worked with Norton Co. and found their 37C46-15V wheel for roughing does a good job and gives a good commercial finish. The wheel has to be dressed often and takes approximately 0.007 in. off the wheel for each 0.001 in. off the work. Dress the wheel often, but do not dress the wheel for the last 0.002 in. This will give a very good finish.



Making an intermediate seam weld in the high-alloy lining of the top head of a wood-pulp digester

WELDABILITY OF WROUGHT HIGH-ALLOY MATERIALS

Welded joints in alloys of nickel-molybdenum, nickel-molybdenum-chromium, cobalt-nickel-iron-tungsten and N-155 meet the minimum tensile requirements for the base metal

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Presented at the Thirty-Fifth AWS National Fall Meeting in Chicago, Nov. 1-5, 1954.

ABSTRACT. The increasing use of high alloys as materials of construction has created a demand for more information on their fabricating characteristics. Much development work has been centered on the joining and forming of alloys capable of resisting corrosion, high-temperatures plus oxidation, stresses and mechanical shocks. It is the procedures for joining, and the properties to be expected of weld joints in these alloys, that will be discussed in detail here.

Tests were made with nickel-molybdenum alloys, nickel-molybdenum-chromium alloys, cobalt-chromium-tungsten-nickel alloys, and N-155, an iron-base alloy with large percentages of chromium, nickel and cobalt. These were welded by the submerged arc, inert-gas-shielded arc and metallic-arc methods. The properties of the weld joints and the weld metal were satisfactory and fell within acceptable limits for alloys of this type. All of the

alloys listed will produce weld joints meeting the minimum tensile requirements for the parent metal.

The alloys tested have characteristics of the austenitic type and, generally speaking, can be welded by all of the methods common to the austenitic stainless steels. Certain of the alloys, however, exhibit characteristics that are peculiar to that alloy composition during welding and fabricating operations.

Nickel-molybdenum alloys, for example, are subject to stress-cracking in the hot-short range and are the most critical of the alloys in this respect. Nickel-molybdenum-chromium and cobalt-chromium-tungsten-nickel alloys are subject to work-hardening during forming operations. These alloys, however, are not as susceptible to stress cracking during welding of restrained components as are the nickel-molybdenum alloys. N-155 alloy is subject to microfissuring if proper welding techniques are not used. All of these alloys can be welded satisfactorily when proper procedures are followed.

Introduction

The introduction of stainless steels—the 12% chromium type—was made approximately 40 years ago. While we were told that samples of chromium-bearing steel were shown at the Paris Exposition in 1900, its most valuable properties apparently went unrecognized for a score of years.[1]

Since then, very rapid metallurgical progress has been made in the fields of corrosion and oxidation-resistant alloys. Many alloy compositions stemming from the stainless steel discovery were developed. As our science and technology advanced, process problems became more complex, and a demand was created for alloys with even greater resistance to corrosion, high temperature plus oxidation, or a combination of corrosion and high temperature plus oxidation.

To meet this demand, a number of metals called "high alloys" were developed. These are still being modified and further tailored for use in specific applications such as heat exchangers, isomerization towers, vats, vessels and tanks used by the petroleum, petrochemical and chemical industries. Then, to meet the requirements of mechanical assemblies operating at high temperatures, the high alloys were designed to operate at high stresses from 1200 to 1900° F.

Since these alloys are tailored for specific applications, they may in a sense be called "specialty alloys," with physical and mechanical properties peculiar to their basic compositions. Therefore, their properties must be thoroughly studied before they can be successfully welded and fabricated.

Much development work has been centered on the joining and forming of high-alloy materials and it is the purpose of this paper to show what we have learned both from the standpoint of welding procedures and the properties to be expected of weld joints.

High alloys may be separated into three groups according to the types of exposure in service to which they are usually subjected:

1. Corrosion.
2. Heat. (Good oxidation resistance and load-carrying ability necessary.)
3. Combined corrosion and heat.

The joining processes that are generally used for stainless steel are applicable to the high alloys, with or without some slight variations. The processes to be discussed are those used currently in the fusion welding of nickel and cobalt-base corrosion and heat-resistant alloys. The data accumulated in the evaluation of the processes, techniques and procedures are (except where noted) based on the use of mill-annealed base material.

Since all of the processes discussed may produce satisfactory joints, the choice of the process is determined by:

1. Alloy grade.
2. Metal gage.
3. Weldment design.
4. Joint design.
5. Joint position.
6. Special problems incurred in erection or fabrication.

Good properties are exhibited in as-welded joints. For maximum corrosion resistance, however, the material should be given a solution-type anneal after welding.

Table 1 lists the nominal chemical compositions and average mechanical properties of the alloys indicated. Similar data for stainless steels are listed for comparative reference.

Table 1—Chemical Compositions and Average Mechanical Properties

Elements	*Stainless steel (347-321)*	*Nickel, molybdenum*	*Nickel, molybdenum, chromium*	*Cobalt, chromium, nickel, tungsten*	*Alloy N-155*
Fe	Balance	4.0–7.0	4.5–7.0	3.00 max	Balance
Cr	18.0	1.00 max	15.5–17.5	19.0–21.0	20.0–22.5
Ni	10.0	Balance	Balance	9.0–11.0	19.0–21.0
C	0.10 max	0.05 max	0.15 max	0.15 max	...
W	...	...	3.75–5.25	14.0–16.0	2.0–3.0
Si	...	1.00 max	1.00 max	1.00 max	...
Mn	...	...	1.00 max	1.0–2.0	...
Co	...	...	...	...	18.5–21.0
Cb	10 × C (347)	...	...	Balance	...
Ti	× C (321)	...	...	...	...
Mo	...	26–30	16–18	...	2.5–3.5
Cb-Ta	...	...	...	...	0.75–1.25
Ultimate strength (annealed), psi	85,000	135,000	130,000	155,000	110,500
Yield strength, 0.2% offset, psi	30,000	60,000	54,000	70,000	71,000
Elongation, %	50	45	35	55	44

Presented at the Thirty-Fifth AWS National Fall Meeting in Chicago, Nov. 1-5, 1954.

erties of the weld joints and the weld metal were satisfactory and fell within acceptable limits for alloys of this type. All of the

Surface Preparation

In all instances, the welding surface and adjacent base metal should be thoroughly cleaned down to bright metal immediately before welding. All grease, oil, crayon marks, dirt, etc., should be removed. If foreign materials are present in the forms mentioned, they may be removed by scrubbing with trichlorethylene or some other suitable solvent material. The surface should be wiped clean before welding.

Joint Design and Edge Preparation

Many types of joints are used in the fabrication of fusion-welded equipment. The type chosen will not necessarily change with a change of welding process since these joint designs are standard. To make these joints suitable for automatic welding operations, such as submerged arc or inert-gas-shielded arc, certain slight modifications may be necessary.

These alloys do not have the fluidity of steel during a welding operation. Therefore, when V or U or J grooves are used, it is necessary to provide a slightly larger clearance than would be needed for steel. This larger clearance allows for cleaning and better accessibility during welding. In general, V joints should be beveled to a 75- to 80-deg included angle, U joints beveled to a 30-deg included angle with a minimum of a $^1/_4$-in. bottom radius and J grooves should have a 15-deg bevel with a minimum bottom radius of $^5/_{16}$-in. Tee joints between dissimilar material thicknesses should have a bevel of 45 deg. See Table 2 for comparison to standard joint types.

The use of backing rings should be avoided when designing containers or pipe lines to be used for carrying corrosive media. Crevices cannot be avoided in this type of design and these may become points of cell-type and stress corrosion, notch effect and root cracking.

In general, material 12-gage and heavier should be beveled and welded from both sides. When joining material of dissimilar thicknesses, the heavier section should always be beveled for ease of welding. For material 12-gage and thinner, the welding may be accomplished from one side by using proper edge spacing to allow full penetration. Care should be exercised to eliminate nonuniform penetration. This condition can leave undesirable crevices and voids in the underside of the joint which, as noted before, contribute to areas of accelerated corrosion. Nonuniform penetration in material used for high-temperature applications creates stress risers for focal points of mechanical failure.

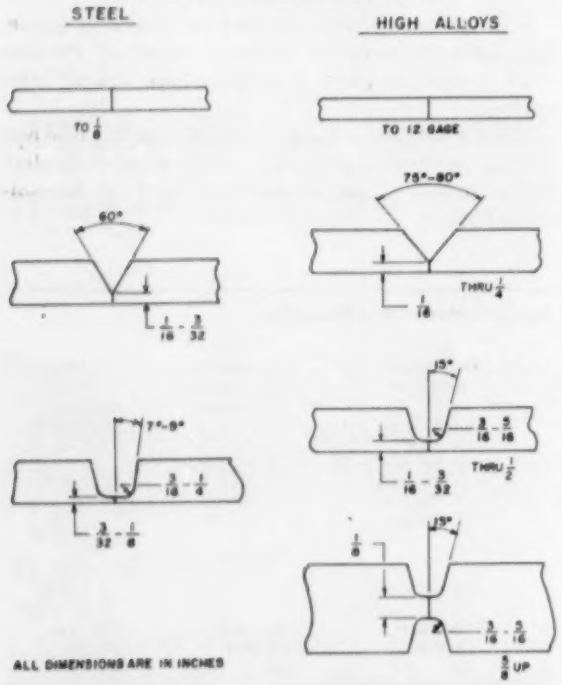
Welding from both sides is recommended wherever possible. When this is not practical, the joint spacing should be increased and a copper backing bar used. Currents slightly higher than normal are then used to obtain complete penetration.

Beveling by machine is the surest way to obtain correct fits. A planer, shaper, grinder or other machine tool can do this job accurately. When sheared sheet or plate is used, the sheared edges should be ground back approximately $^1/_{16}$ in. to remove any stressed material before the edge is prepared for welding. In *all* instances, the edges should be squared, *aligned properly*, and tacked before welding. The alignment and edge preparation are particularly important in this class of alloys. Any misalignment causes variation in bead contour, gap width and stresses in the weld area. These factors contribute to cracking in the weld joint. Since the initial metal cost is fairly high, the slight additional cost of careful preparation to assure good welds is well justified. Thermal cutting and beveling of plates can be done, but these are not recommended procedures.

In general, the V joint preparation is used for butt welds in plate thicknesses up to $^1/_4$ in. and a U joint for greater thicknesses. The V or U joint is used where the welded material will be exposed to high stress. When these types of weld construction are used, the stress will act axially. The lap or tee joint may be used for conditions of lower stress.

The U joint preparation is preferred for material greater than $^1/_4$ in. in thickness. While the cost of preparation may be increased by this type of joint, the amount of welding materials and man-hours needed for welding will be much less than if a V joint is used. Also, the amount of residual stress will be lower since less weld material is required and less transverse shrinkage is incurred.

Table 2—Standard Weld Joints



Jigs and Fixtures

The use of jigs and fixtures for assembly and subsequent welding provide definite economies in reduced

Elongation, %	50	45	35	55	44

welding cost, greater speed in welding and ease of producing welds of uniformly high quality. Proper jigging and clamping of the weld joint makes the welding operation easier, and holds buckling and warping to a minimum. It is generally desirable to provide a grooved backing bar of some sort. This serves as a chill to the base material and as a support to excessive hang-through of the weld bead. The use of a backing bar contributes to the ease of welding and helps to obtain a more uniform bead penetration.

Steel or cast iron may be used in fixtures for oxy-acetylene welding. When the arc process is used, the portion of the fixture contacted by the arc should be copper.

When using a backing or chill bar, a groove of the proper contour is usually provided to permit good penetration and bead contour. For oxy-acetylene welding, the backing bar should be grooved to $^1/_{16}$ in. deep and about $^5/_{16}$ in. wide. The groove width and hold-down spacing should be adjusted to obtain the best conditions of heat transfer, plate hold-down and ease of welding. For arc welding, the grooves should be of a minimum depth, usually from $^3/_{32}$ to $^1/_{16}$ in., and approximately $^3/_{16}$ in. wide. The corners of the groove should be rounded. Square corners cause poor bead contour, flux pockets, and nonuniform heat transfer. Jigs and fixtures can be used to particular advantage when using the inert-gas-shielded arc process.

Nickel-Molybdenum

This is a corrosion-resistant alloy designed to handle boiling hydrochloric acid, wet hydrogen chloride, sulfuric acid and other corrosives. It is readily weldable in the horizontal and flat positions, and may be welded in the vertical position if sufficient skill is acquired. However, welding in the vertical position is not recommended except in cases of emergency. The alloy is very fluid during deposition, and careful handling is required to avoid undercutting and poor bead contour. Welding techniques in general are similar to those used in the welding of stainless steel. However, the following precautions should be observed:

1. Use stringer beads at all times (no weaving).
2. Do not preheat or postheat.
3. Keep weldment as cool as possible. Quench with rag and water wherever possible.
4. Use striking tabs where possible.

By observing these rules and using proper joint design (refer to Table 2), few difficulties will be experienced. A study of the mechanical properties of the alloy before designing the weldment will be helpful, too (see Fig. 1). Notice that the elongation drops from 40% at 800° F to 9% at 1350° F, and then increases to approximately 30% at 1800° F. The reduction of area (Fig. 2) follows this same trend. These data represent the characteristics of the alloy, and are directly related to the way it performs during welding and subsequent cooling cycles. From these data, it may be

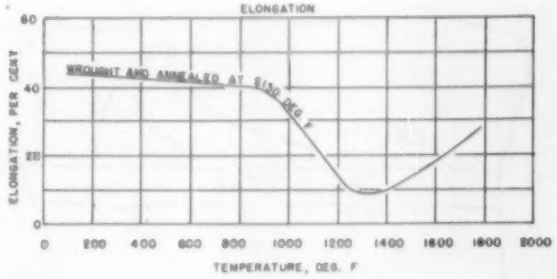


Fig. 1 Elongation, nickel-molybdenum alloy

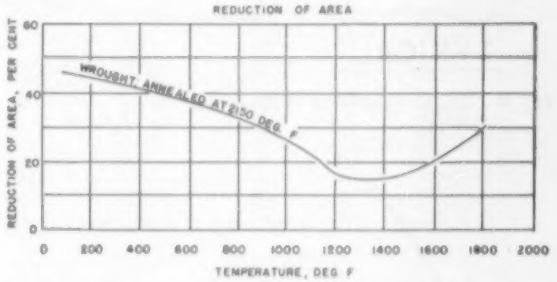


Fig. 2 Reduction of area, nickel-molybdenum alloy

Fig. 3 Cracking in heat-affected zone of nickel-molybdenum alloy sheet welded by the inert-gas-shielded arc method. × 100

seen that it is important to pass through the temperature range of 1800 to 1100° F as rapidly as possible during cooling, and to design and position the welds in a manner that will subject them to the least restraint.

Cracking is rarely experienced in the weld deposit. When cracking does occur, it usually appears in the base material (Fig. 3). Such cracking is very fine, approximately $^1/_8$ in. long, and usually transverse to

quent welding provide definite economies in reduced

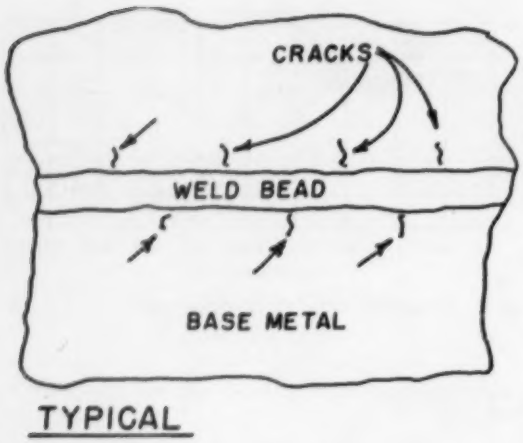


Fig. 4 Cracks transverse to the weld

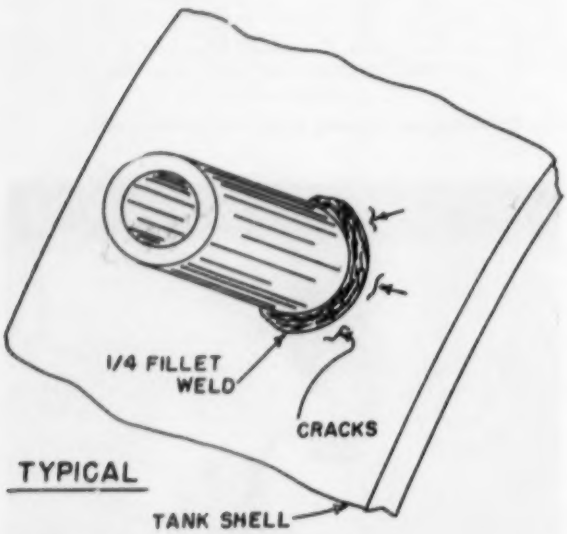


Fig. 5 Cracks parallel to the weld

the weld (Fig. 4). In cases of extreme restraint, where the welded part might be a spud in a tank shell, for example, the cracking will occur parallel to the weld (Fig. 5). However, when the characteristics of this alloy are understood, little or no difficulty is experienced in its welding.

The causes behind these two types of cracking, transverse or parallel to the weld, may be explained as follows:[2]

1. Cracks in the parent metal, perpendicular to the weld, occur during the deposition of the electrode into the weld joint. At this time the parent metal is in the process of expanding as a result of heat absorption from the hotter weld deposit. When the parent metal reaches the hot-short range, stresses set up by this expansion cause small cracks to propagate from the edge of the deposit through the heat-affected zone, which is approximately $^3/_{32}$ in. wide (Figs. 3 and 4).

2. When fillet welding spuds or other fittings into a tank, or welding other parts having restraint, cracks may appear approximately $^1/_8$ in. from the weld and longitudinal to the weld bead. This cracking is caused by shrinkage across the weld fillet face and subsequent cracking in the heat-affected zone (Fig. 5).

Inert-Gas-Shielded Arc Welding

All the recommendations discussed up until now should be followed when the inert-gas-shielded arc process is used for joining. In general, the conditions observed while welding the high-alloy materials by this method were similar to those which occur during the welding of stainless and high-alloy steels.

The preferred power supply is direct-current, straight polarity with the electrode at negative potential. A tungsten electrode of the smallest diameter to carry this current is recommended, and the gas may be either pure argon, argon with 1% oxygen or pure helium. The gas flow of argon should be about 30 cfs. Electrodes are normally operated at maximum current density in order to obtain the best arc stability.

Table 3—General Welding Details for Manual Inert-Gas-Shielded Arc Method for Nickel-Molybdenum and Nickel-Molybdenum-Chromium Alloys

Base metal thickness, in.	*Type of weld*	*Number of beads**	*Amperes*		*Tungsten electrode diam, in.*	*Filler rod diam, in.*	*Argon flow, lpm*
			Flat	*Vertical (up)*			
$^1/_{16}$	Butt	1	60–90	55–75	$^1/_{16}$–$^3/_{32}$	$^1/_{16}$–$^3/_{32}$	10–12
	Fillet	1					
	Corner	1					
$^3/_{32}$–$^1/_8$	Butt	1–2	100–125	80–110	$^3/_{32}$	$^1/_{16}$–$^3/_{32}$–$^1/_8$	10–12
	Fillet	1					
	Corner	1					
$^3/_8$	Butt	3	140–200	110–160	$^1/_8$–$^3/_{32}$	$^3/_{32}$–$^1/_8$–$^5/_{32}$–$^3/_{16}$	16–18
	Fillet	2					
	Corner	2					
$^1/_2$	Butt	5	200–325	160–260	$^5/_{32}$–$^3/_{16}$ $^1/_8$	$^1/_8$–$^3/_{32}$–$^3/_{16}$	18–20
	Fillet	4					
	Corner	3					

* Current requirements—DCSP.

REMARKS: (1) These data are general: The number of beads and filler rod diameter may be varied to account for setup and operator variables. (2) Stringer beads should be deposited on welds $^1/_8$ in. and up. (3) A more than normal tungsten projection will be required for fillet welds and first few beads on heavy L. (4) Smaller diameter filler rods and more beads may be employed on position welds for adequate weld metal control. (5) If reverse sides of butt welds do not show adequate penetration, they should be ground to sound metal and a backing bead made. (6) Protection should be provided on reverse side of welds where possible.

Table 4—Mechanical Properties of Nickel-Molybdenum Alloy in the As-Welded Condition

Bar No.	*Weld method*	*Specimen thickness, in.*	*Ultimate strength, psi*	*Yield strength, 0.2% offset*	*Elongation in 2 in., %*	*Angle of deformation after 180-deg transverse bend*	*Weld failure*	*Hardness-Rockwell A (average)* *Plate*	*By-weld*	*In-weld*
1-O	Metallic arc	1/8	121,600	67,500	16.0	166	No	61.8	61.2	52.8
2-P	Metallic arc	1/8	123,000	66,500	20.0	167	No	61.8	61.2	52.8
3-P	Metallic arc	1/8	118,000	69,300	15.0	167	No	61.8	61.2	52.8
1-B	Inert-gas-shielded arc	1/8	117,200	58,800	30.0	172	No	56.8	58.8	54.6
2-B	Inert-gas-shielded arc	1/8	121,600	57,600	31.0	172	No	56.8	58.8	54.6
3-B	Inert-gas-shielded arc	1/8	116,000	56,600	29.0	172	No	56.8	58.8	54.6
3-D	Submerged arc	1/4	112,000	67,000	19.0	30	Yes			
4-D	Submerged arc	1/4	116,000	70,500	20.0	30	Yes			
5-C	Submerged arc	1/4	114,000	65,800	22.5	30	Yes			
								Rockwell B		
A3M2	Metallic arc	1/4	123,900	55,500	40.0	O.K.	No	91	92	93
A3M3	Metallic arc	1/4	123,000	52,800	42.0	O.K.	No	91	92	93
A3M4	Metallic arc	1/4	124,000	53,800	45.0	O.K.	No	91	92	93
A3M6	Inert-gas-shielded arc	1/4	115,000	60,100	30.0	O.K.	No	92	93	95
A3M3	Inert-gas-shielded arc	1/4	118,200	59,800	30.0	O.K.	No	92	93	95
A3M5	Inert-gas-shielded arc	1/4	112,000	58,800	20.0	O.K.	No	92	93	95
A4M7	Metallic arc	3/8	120,000	61,100	30.0	O.K.	No	97	97	93
A4M5	Metallic arc	3/8	125,600	53,400	45.0	O.K.	No	97	97	93
A4M3	Metallic arc	3/8	121,100	59,600	37.0	O.K.	No	97	97	93
A4M3	Inert-gas-shielded arc	3/8	124,900	48,500	42.0	O.K.	No	92	94	90
A4M1	Inert-gas-shielded arc	3/8	119,500	54,400	32.0	O.K.	No	92	94	90
A4M3	Inert-gas-shielded arc	3/8	127,900	60,900	34.0	O.K.	No	92	94	90
	Minimum mechanical properties of fully annealed base material (ASME Boiler Code, U-68, Section VIII, Case 1173)									
1		To 3/16	115,000	48,000	45.0					
2		To 3/4	100,000	45,000	40.0					
3		To 1 1/2	90,000	45,000	35.0					

It has been found, however, that the use of 1/16-in. diam. electrodes under these conditions creates excessive electrode contamination and results in excessive tungsten deposits in the weld puddle. This problem is materially reduced by using a 3/32-in. diam electrode ground to a long needle-like point.[3]

See Table 3 for general details. Precautions should be taken to prevent drafts from dispersing the gaseous shield. If the shield is dispersed, oxidation and subsequent porosity may result. It has been found, too, that excess nitrogen will cause porosity in inert-gas-shielded arc welds of this alloy. This condition can be minimized by shielding the reverse side of the weld. Any aspiration of air into the weld deposit can be eliminated by using the proper joint design and adequate shielding of the weld. It is also practical to use some form of backing paste to assure root protection as well as to secure good underbead contour.[4]

Submerged Arc Welding

Nickel-molybdenum alloys may be welded successfully by the submerged arc process. For best welding conditions, a power supply of direct-current, reverse polarity (electrode positive) should be used. In general, weld strengths in excess of 105,000 psi, yield strength of 60,000 psi and elongations of 18–22% can be attained by this process.

A granular welding composition should be selected to give maximum protection to the welds and to keep silicon and manganese pickup at a minimum. These impurities will impair the corrosion-resistant properties of this alloy. The heat-affected zone will be wider with the use of this process, and the cracking tendency may be increased by the higher heat input. However, this condition can be offset, to a large extent, by careful joint design. See Table 4 for some weld properties of this material.

It is apparent from these data that the ductility of submerged arc welds is rather poor and subsequent transverse bead tests were very unsuccessful because of weld cracking with bend angles of 60 to 90 deg. Figures 6 and 7 show the material in the fully annealed condition and in the as-welded condition.

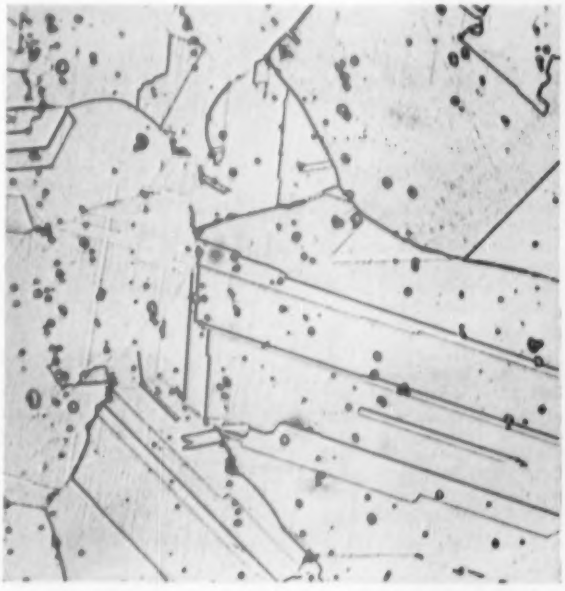


Fig. 6 Nickel-molybdenum alloy sheet solution annealed. × 250

and a backing bead made. (6) Protection should be provided on reverse side of welds where possible.

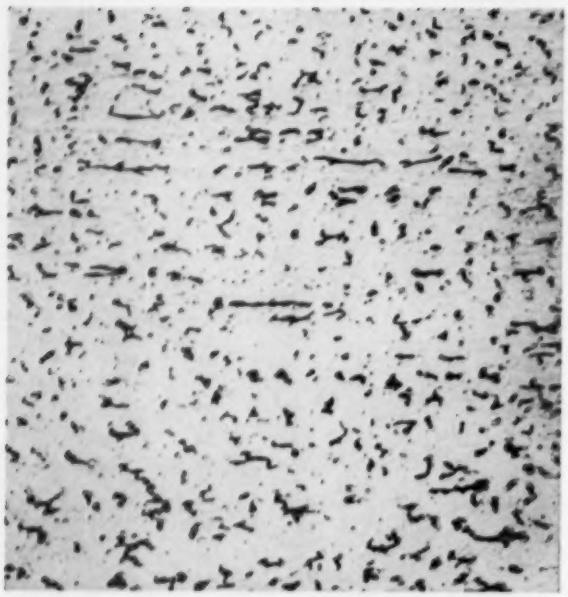


Fig. 7 Nickel-molybdenum alloy as-welded by the inert-gas-shielded arc method. X 250

Nickel-Molybdenum-Chromium

This is a corrosion-resistant alloy designed to withstand strong oxidizing media such as solutions of chlorine and hypo-chlorite, wet chlorine gas and acid solutions containing strongly oxidizing salts. In general, it may be welded and fabricated using the same procedures and techniques as applied to the nickel-molybdenum alloy. This alloy has slightly lower tensile strength, yield strength, and elongation (See Table 1). However, upon examination of the elongation vs. temperature diagram (Fig. 8), it can be determined that the nickel-molybdenum-chromium alloy is not subject to as much loss of ductility in the range of 900 to 1800° F as the nickel-molybdenum alloy, and, therefore, is not as prone to hot-short cracking. This characteristic allows more latitude in the welding and fabricating of this alloy; however, the welding procedures described under nickel-molybdenum alloy should be followed for ease of welding. When cracking does occur because of severe restraint, it usually follows the same pattern as described for nickel-molybdenum alloy and as shown in Figs. 3, 4 and 5.

For metallic-arc joining of this material, a direct current reverse polarity power source is used.

To obtain the ultimate mechanical and corrosion properties, lime-coated electrodes should be used. Although this material is readily welded with lime-titania type electrodes, it is generally accepted by the field that better metallurgical properties and better corrosion resistance of welds are obtained by the use of lime-coated electrodes.

Inert-Gas-Shielded Arc Welding

Nickel-molybdenum-chromium alloy is not as prone

Table 5—Mechanical Properties of Nickel-Molybdenum-Chromium-Alloy in the As-Welded Condition

Bar No.	*Weld method*	*Specimen thickness, in.*	*Ultimate strength, psi*	*Yield strength, 0.2% offset*	*Elongation in 2 in., %*	*Angle of deformation after 180-deg transverse bend*	*Weld failure*	*Hardness-Rockwell A (average)* *Plate*	*By-weld*	*In-weld*
1J	Metallic arc	1/8	118,200	67,200	20.0	169	No	58.7	59.5	60.2
2J	Metallic arc	1/8	115,000	68,600	17.5	168	No	58.7	59.5	60.2
3J	Metallic arc	1/8	116,900	68,300	19.5	167	No	58.7	59.5	60.2
6I	Inert-gas-shielded arc	1/8	124,000	67,000	24.0	176	No	59.8	56.5	59.5
7I	Inert-gas-shielded arc	1/8	121,000	66,100	25.0	178	No	59.8	56.5	59.5
8I	Inert-gas-shielded arc	1/8	120,000	65,600	22.5	177	No	59.8	56.5	59.5
								Rockwell B		
B3M7	Metallic arc	1/4	128,500	66,500	28.0	O.K.	No	87	90	91
B3M2	Metallic arc	1/4	132,000	64,000	35.0	O.K.	No	87	90	91
B3K3	Metallic arc	1/4	127,100	63,500	28.0	O.K.	No	87	90	91
B3M6	Inert-gas-shielded arc	1/4	109,000	58,200	23.0	O.K.	No	97	97	96
B3M3	Inert-gas-shielded arc	1/4	113,200	58,900	32.0	O.K.	No	97	97	96
B3M5	Inert-gas-shielded arc	1/4	117,800	58,000	32.5	O.K.	No	97	97	96
1B	Submerged arc	1/4	104,500	68,000	13.0					
2B	Submerged arc	1/4	99,000	69,000	12.5					
3B	Submerged arc	1/4	106,000	72,000	12.5					
								Rockwell A		
B4M3	Metallic arc	3/8	123,000	64,600	35.5	O.K.	No	51.8	51.3	53.6
B4M6	Metallic arc	3/8	118,000	65,900	22.0	O.K.	No	51.8	51.3	53.6
B4M4	Metallic arc	3/8	110,000	67,300	16.5	O.K.	No	51.8	51.3	53.6
B4M7	Inert-gas-shielded arc	3/8	112,700	64,300	22.0	O.K.	No	54	56.2	55.6
B4M2	Inert-gas-shielded arc	3/8	118,000	64,300	29.0	O.K.	No	54	56.2	55.6
B4M6	Inert-gas-shielded arc	3/8	116,500	64,000	25.0	O.K.	No	54	56.2	55.6
1F	Submerged arc	1/2	65,000*	...	...					
2F	Submerged arc	1/2	101,500	63,800	12.5					
3F	Submerged arc	1/2	89,000*	...	...					
			Minimum mechanical properties of fully annealed base material							
		To 3/16	115,000	50,000	40					
		To 3/4	100,000	45,000	30					
		To 1 1/2	90,000	45,000	20					

* Imperfections in weld (visible).

to porosity when joined by this process as is the nickel-molybdenum alloy and, in general, exhibits slightly better welding characteristics. This property is attributed to the added chromium content of this alloy. The same welding practice should be adhered to as for nickel-molybdenum alloy (Table 3).

Table 5 shows the mechanical, as-welded properties of nickel-molybdenum-chromium alloy. A comparison of the properties listed in Tables 4 and 5 shows that the nickel-molybdenum-chromium alloy exhibits slightly lower tensile strength, yield strength and elongation. This is in accord with the properties of the parent materials (Table 1).

Inert-gas-shielded arc welding has been generally accepted as the best method of joining materials to be used in corrosion applications. Welds produced by this method are of slightly higher quality than those obtained with the metallic-arc method. Possible slag inclusions, pickup of stray elements from the coating, excessive burn-out of metallic elements in the rod, etc., are not as pronounced. However, the time element may be the deciding factor, in which case the metallic-arc method would, of necessity, be used.

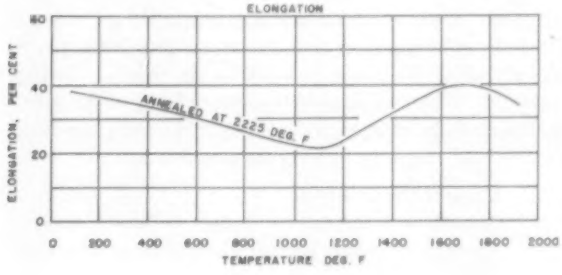


Fig. 8 Elongation and reduction of area, nickel-chromium-molybdenum alloy (sheet, 0.065 in. thick)

Submerged Arc Welding

Nickel-molybdenum-chromium alloy may be welded by the submerged-arc welding process under the same conditions used for welding nickel-molybdenum alloy. The comparison shown in Tables 4 and 5 indicates that a lower level of mechanical properties exists for the nickel-molybdenum-chromium alloy, however. Submerged arc welds in the nickel-molybdenum-chromium alloy exhibit the same poor transverse bend properties as described in the previous discussion of nickel-molybdenum alloy.

Figures 9 and 10 compare solution annealed material and as-welded material. Figures 11 and 12 show some bend-test samples of as-welded material of nickel-molybdenum and nickel-molybdenum-chromium.

N-155 Alloy

This material is used in parts exposed to high tem-

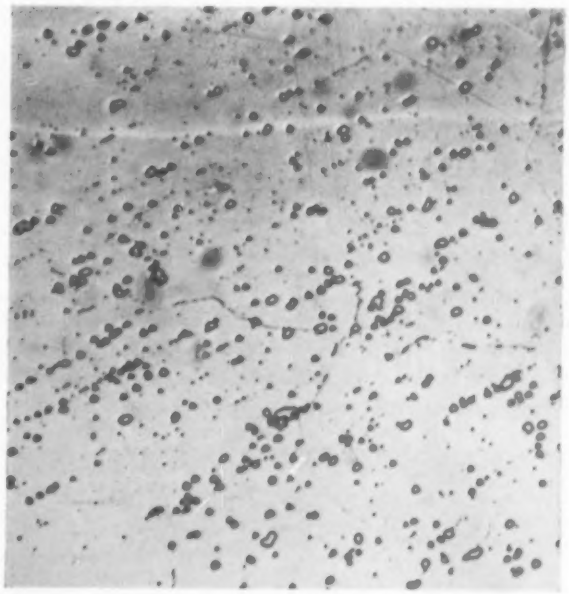


Fig. 9 Nickel-molybdenum-chromium alloy sheet, solution annealed. × 250

Fig. 10 Nickel-molybdenum-chromium alloy as-welded by the inert-gas-shielded arc method. × 250

perature and oxidation. Some typical uses include gas-turbine engine tail cones, cabin hot-air heat exchangers, after-burner sections, etc. The alloy is subject to the same welding and fabricating procedures as outlined in the foregoing sections, although the difficulties encountered are of a different nature. This material has a slightly lower tensile strength, yield strength, and elongation than the nickel-molybdenum and nickel-molybdenum-chromium alloys (Table 1).

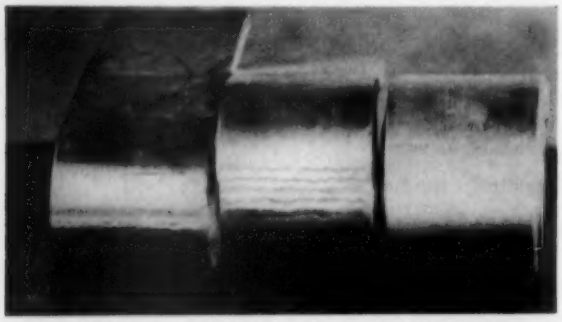


Fig. 11 Nickel-molybdenum alloy bend test samples of different thicknesses. *Left to right:* 0.125 in., 0.250 in., 0.375 in.

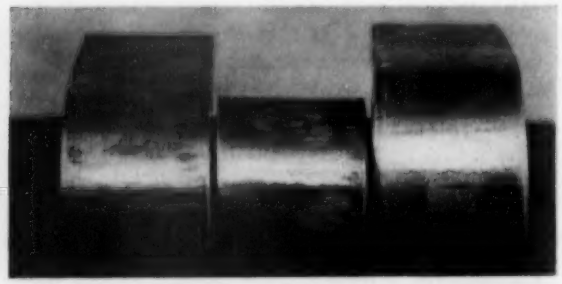


Fig. 12 Nickel-molybdenum-chromium bend test samples of different thicknesses. *Left to right:* 0.125 in., 0.187 in., 0.250 in.

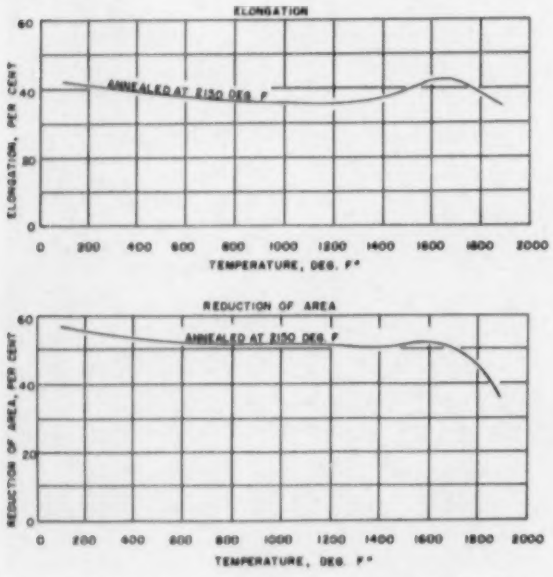


Fig. 13 Elongation and reduction of area, N-155 alloy wrought (sheet, 0.052 in. thick)

On examination of the elongation vs. temperature diagram (Fig. 13), it is noted that this alloy does not have as critical a hot-short range as the other alloys thus far discussed, but retains nearly a straight-line function to about 1700° F. The ductility drops off sharply at this point. This phenomenon appears to correlate with the sensitivity of this alloy to microfissuring when the proper welding procedures are not pursued. Since the ductility is so markedly reduced at 1700° F, it is assumed that microcracking takes place directly after the liquid to solid transformation during welding. When welding this alloy (with any process), therefore it is extremely important to pass through this critical range as quickly as possible by:

1. Using stringer beads.
2. Quench welding (use wet rags or water spray).
3. Keeping interpass temperature at room temperature.
4. Keep base material at room temperature.

The nature of the cracking or microfissuring encountered in welding N-155 is very similar to that encountered in chromium-nickel stainless weld metal (see Fig. 14). Cracking is always more prevalent in the fully austenitic weld metals than in the partially ferritic weld metals.[5]

The influence of the electrode coating is mainly one of degree—that is, if other conditions favor cracking, the type of coating may increase or decrease the amount of cracking. For example, an alloy which is notably crack-sensitive when deposited by titania type of electrode, may become free from cracks by eliminating the titania from the coating.[5] This conclusion applies to N-155 alloy and has been proved by experimentation using the standard circular groove type specimen to indicate weld bead cracking. Electrodes furnished for this material are, therefore, lime coated unless otherwise specified.

It has often been conjectured in various papers that columbium attributes crack sensitivity to a material; however, this has neither been proved nor disproved for N-155 alloy. (See Table 6 for some as-welded mechanical properties.)

Inert-Gas-Shielded Arc Welding

N-155 is welded readily by the inert-gas-shielded arc method using the previously outlined procedures and practices (Table 3). For as-welded mechanical properties, see Table 6. A minimum of heat input should be used, followed by rapid cooling of the weld deposits

Fig. 14 Cracking or microfissuring encountered in welding N-155 alloy

Table 6—Mechanical Properties of N-155 Alloy in the As-Welded Condition

Bar No.	*Weld method*	*Specimen thickness, in.*	*Ultimate strength, psi*	*Yield strength, 0.2% offset*	*Elongation in 2 in., %*	*Angle of deformation after 180-deg transverse bend*	*Weld failure*	*Hardness (Rockwell B) Weld*	*Hardness (Rockwell B) Adjacent to weld*
C1H1	Inert-gas-shielded arc	1/8	108,000	61,000	24.0	O.K.	No	91	94–89
C1H3	Inert-gas-shielded arc	1/8	108,900	61,000	22.5	O.K.	No	91	94–89
C1H4	Inert-gas-shielded arc	1/8	107,300	63,300	21.5	O.K.	No	91	94–89
C1H5	Inert-gas-shielded arc	1/8	110,900	...	23.0	O.K.	No	91	94–89
C1H7	Inert-gas-shielded arc	1/8	105,800	60,900	20.0	O.K.	No	91	94–89
C1M1	Metallic arc	1/8	115,900	60,500	27.0				
C1M2	Metallic arc	1/8	117,000	62,200	28.0				
C1M3	Metallic arc	1/8	115,300	60,100	27.0				
C4H2	Inert-gas-shielded arc	1/4	114,700	66,000	23.0	O.K.	No	90	91–91
C4H3	Inert-gas-shielded arc	1/4	114,900	64,500	24.0	O.K.	No	90	91–91
C4H4	Inert-gas-shielded arc	1/4	107,000	64,500	17.5	Failed 60 deg	No	90	91–91
C4H7	Inert-gas-shielded arc	1/4	109,100	65,100	18.0	O.K.	No	90	91–91
C4M1	Inert-gas-shielded arc	3/8	97,000	61,500	14.0	Failed 60 deg	No	90	91–91
C4H3	Inert-gas-shielded arc	3/8	108,900	57,500	18.5	O.K.	No	90	93–94
C4H5	Inert-gas-shielded arc	3/8	117,500	63,000	28.5	O.K.	No	90	93–94
C4H7	Inert-gas-shielded arc	3/8	100,500	60,700	14.0	O.K.	No	90	93–94
C4M3	Metallic arc	3/8	99,300	66,200	27.0	O.K.	No	92	90–90
C4M4	Metallic arc	3/8	117,900	64,800	30.5	O.K.	No	92	90–90
C4M5	Metallic arc	3/8	114,900	64,300	26.0	O.K.	No	92	90–90
C4M6	Metallic arc	3/8	118,100	66,300	30.0	O.K.	No	92	90–90
			Minimum mechanical properties of fully annealed base material						
		To 3/16	100,000	70,000	40.0				

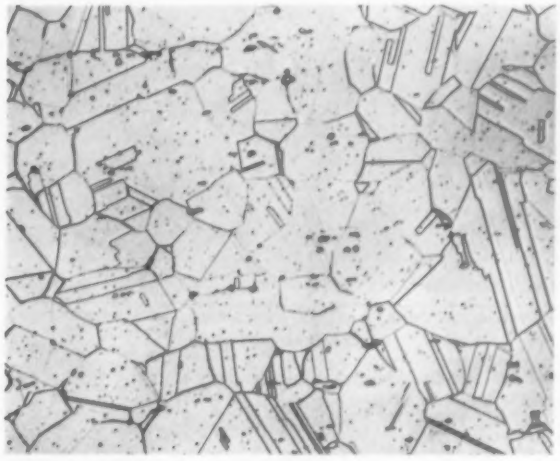


Fig. 15 N-155 alloy sheet solution annealed. X 250

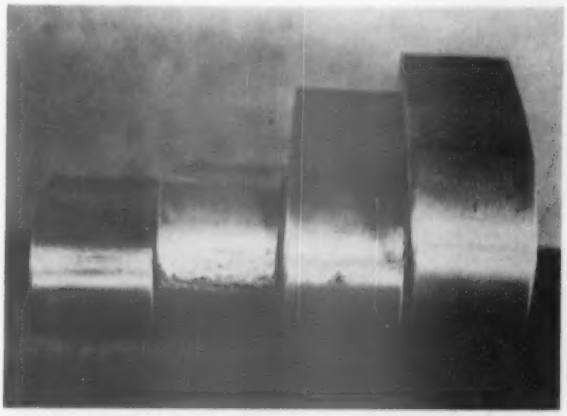


Fig. 17 N-155 alloy bend-test samples of different thickness. *Left to Right:* 0.125 in., 0.250 in., 0.375 in. and 0.500

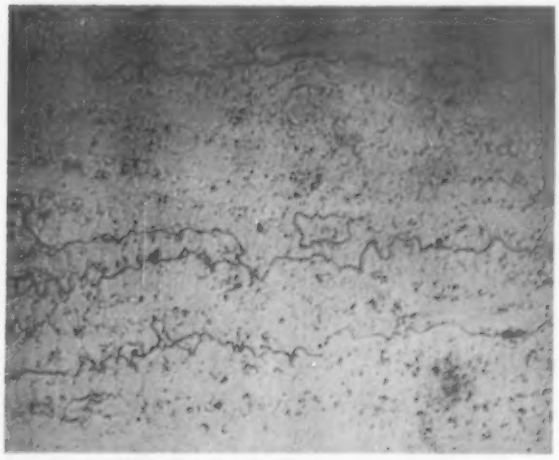


Fig. 16 N-155 alloy as welded by metallic arc. X 250

Submerged Arc Welding

This material does not lend itself readily to submerged arc welding of heavy plate, because of the microfissuring characteristic. In general, submerged arc welds in heavy plate produce high heat input to the base material and slow cooling. Welds may be made satisfactorily in the thinner gages, with this process. Figures 15 and 16 compare the solution-annealed material with as-welded material. Figure 17 shows some bend-test samples of as-welded N-155 alloy in different thicknesses.

Cobalt-Chromium-Tungsten-Nickel Alloy

This material is a high-temperature oxidation-resistant alloy that has high strength and oxidation resistance at 1800° F. It possesses higher mechanical properties, such as tensile strength, yield strength and elon-

gation, than those shown by the other alloys discussed (Table 1). This material is readily weldable by the several processes discussed in this paper and by utilizing the same procedures, joint design, etc., no difficulties should be encountered.

Figure 18, the elongation vs. temperature diagram shows that the elongation drops off sharply at 900 and at 1300° F where it is approximately 15%, and rises to 20% at about 1900° F. It is subject to hot-short cracking, much the same as nickel-molybdenum alloy, and the same considerations used for the nickel-molybdenum alloy should be applied to it to secure good welds.

Cobalt-chromium-tungsten-nickel alloy is readily fabricated. Since it has high yield strength, however, it is necessary to have forming equipment of adequate capacity. Figures 19 and 20 compare the solution annealed material with as-welded material.

Conclusions

Our experience with these high-alloy materials has been that they can be welded, but it is necessary to observe the following general conditions to produce satisfactory weldments:

1. To have a minimum of weld restraint.
2. Keep the base material at approximately room temperature.
3. To maintain good alignment of joints.
4. Use stringer beads.

By the application of these basic principles, the welding problems encountered with high-alloy materials will be minimized.

Acknowledgments

The author is grateful to the many individuals who contributed generously of their time to make possible a compilation of experience in this field. The helpful work of L. E. Stark, W. Morton, Dr. H. R. Spendelow, Jr., and others of the Metals Research Laboratories of Union Carbide and Carbon Corp., and the personnel of the Haynes Stellite Development and Technical Services Department is gratefully acknowledged. The author wishes to acknowledge also the help of his associate, G. A. Fritzlen, for his review and assistance in the preparation of this paper.

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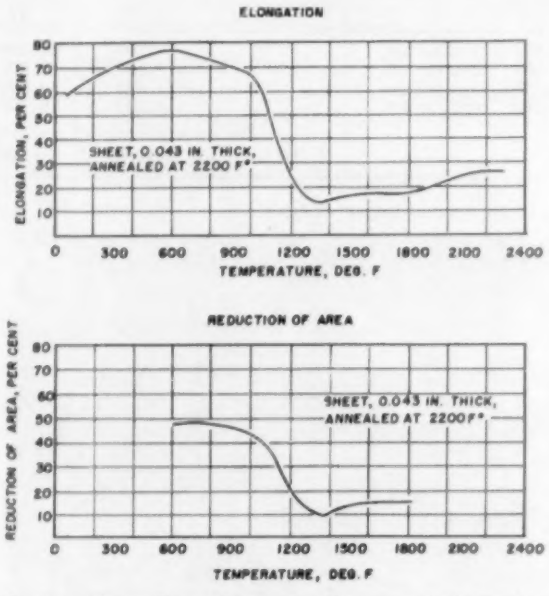

Fig. 18 Elongation and reduction of area, cobalt-tungsten-nickel-chromium alloy

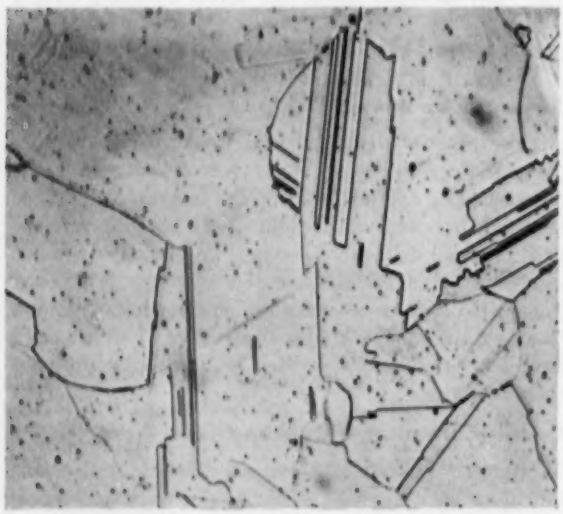

Fig. 19 Cobalt-chromium-tungsten-nickel alloy solution annealed. × 250

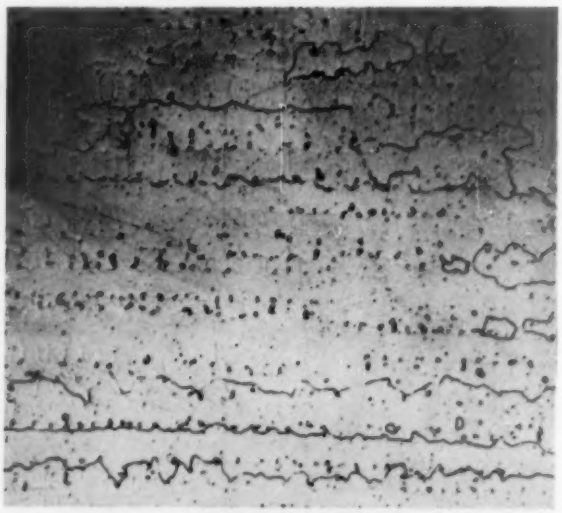

Fig. 20 Cobalt-chromium-tungsten-nickel alloy as-welded by metallic arc. × 250

Installing the selenium rectifier unit in a rectifier-type arc welder

METALLIC RECTIFIERS FOR ARC WELDERS

Rectifiers for use in d-c welders pose some rather peculiar requirements that are unlike other applications. Author gives an analysis of existing rectifiers, outlines the general requirements, and discusses the improvements to be anticipated

By G. K. WILLECKE

The idea of producing d-c weld power by converting alternating current to direct current with rectifiers is not new, although it has been only during the past few years that rectifier welders have appeared on the market.

A rectifier for use in welding service must comply with some rather severe requirements that generally are not met in other applications. An analysis of available rectifiers shows that selenium rectifiers are the best choice for welding service at the present time.

The general requirements for a welding power supply rectifier are developed in terms of electrical and physical characteristics and service conditions.

Improvements to be anticipated in rectifier design are discussed with an evaluation of the effect on welding machine design.

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Presented at the Thirty-Fifth AWS National Fall Meeting held in Chicago, Ill., Nov. 1–5, 1954.

Rectifier Function

As long as certain arc welding requirements call for direct current, and power distribution is based on alternating current, we will be faced with the problem of converting alternating current to direct current.

To accomplish this, we may convert the a-c electrical energy into some other form of energy that can be reconverted readily into d-c electrical energy. The classic and simple way of accomplishing this is to use mechanical energy as the intermediate step. A motor converts the electrical energy into mechanical energy and a d-c generator reconverts the energy back into the electrical form. No doubt, more complex conversion schemes would work but would be of questionable practicability.

A look at the basic fundamentals of the problem immediately points out the fact that any energy conversion process is always less than 100% efficient. This brings about the conclusion that rectifiers are definitely the most clean-cut approach to the problem.

Before an intelligent choice of rectifiers can be made, the weld power requirements should be reviewed. We conclude that the welder must furnish direct currents on the order of magnitude of hundreds of amperes at voltages which may fall anywhere between 20 and 40 v, while the open circuit voltage may go as high as 120 v. The duty cycle of the equipment must be comparable to other equipment available. All welding equipment must be ruggedly designed to withstand hard shop abuse. Cost figures must be low enough to allow the equipment to hold its own in a highly competitive market.

On the basis of current carrying ability only two classes of rectifiers can be accorded any consideration. The two are the group of vapor-arc rectifiers commonly known as ignitrons and thyratrons—and the metallic rectifiers, cuprous-oxide, magnesium-cupric sulfide, selenium and perhaps germanium.

The reason why vapor-arc rectifiers have never been commercially practicable for arc welders is the fact that they operate most effectively at high voltages. In spot and seam welders, where capacitor discharge is used and the rectifiers are used to charge the capacitors to high energy levels, they have found ready acceptance. Their use in arc welders at relatively low voltages has all the aspects of installing a high-power aircraft engine in the family automobile.

Rectifier Requirements

Metallic rectifiers appear to be almost ideally suited for arc welding applications, with the top position presently being held by the selenium rectifier. The reason is the higher voltage rating of the selenium cell compared to the copper oxide and the sulfide types. Metallic rectifier cells carry a voltage rating which corresponds to the highest a-c rms voltage that can be safely applied to the cell. This voltage is dependent upon the manufacturing techniques utilized in building the cell and frequently has represented a compromise between various parameters. Improvements in the manufacturing techniques have yielded an appreciable gain in selenium cell voltage rating from a standard of 18 v around 1945 to approximately 33 v at the present date. Progress is continuing along this line and it is reasonable to expect that reliable higher voltage selenium cells will become available in the future.

The significance of the cell voltage rating becomes apparent when we examine the circuit that is used in welding applications. Although single phase circuits can be used, the economics of the situation dictate the almost exclusive use of a 3-phase bridge rectifier circuit. This means simply that there will be six half-wave rectifying *elements* in the circuit, each carrying one-third of the total output current and each one being subjected to a *peak* inverse voltage of 1.05 times the d-c output voltage. The rms a-c input voltage in this type of circuit then comes out to approximately 74% of the d-c output voltage, assuming sine wave current. Deviations from the norm can be expected if the input wave is not a sine wave.

This, then, establishes the design parameters required for the rectifier welder. If we take, as an example, a 300-amp NEMA* rated welder with an open-circuit voltage of 75 v, we can establish quite positive rectifier requirements.

Voltage Rating

Consider voltage first of all. Depending upon the input voltage wave shape, we can calculate that the "voltage rating" of the stack unit must be in the vicinity of 56 v. If we intend to utilize cells that have a rating of 26 v, we find that two cells in series do not give us the required 56 v so we must go to three cells in series. This gives us a 78-v stack rating which is too high from the economic standpoint. We therefore need at least a "28-v cell" in order to use the two-cell series connection. Ratings above this figure would provide additional safety margin. Ideally, the solution is a "56 volt-or-better" selenium cell which then would reduce the cell requirements to a minimum of one cell per leg.

We might say then that the critical points for the voltage rating of a selenium rectifier cell for satandard design welder are given by the following equation: $E = V/n$, where E is the rated cell voltage, V the a-c input voltage for the required d-c output voltage, and n is a whole number representing the number of cells in series.

In the example previously given, where 56-v a-c input is required, the values would be as follows:

n	E
1	56
2	28
3	18.7
4	14

As an interesting side light we might point out that this is the reason why cuprous-oxide metallic rectifiers with an E value of 4 to 7 v and magnesium-cupric-sulfide with a value of 4 to 5 v are not a very economical

* National Electrical Manufacturers Association.

choice for welder rectifier service, since it would require too many cells in series for adequate voltage rating.

Current Rating

While the proper choice of the number of cells with respect to voltage is a simple and straightforward matter, the current rating presents many complex, conflicting problems.

It should be pointed out that while rectifier applications for such functions as battery charging, electroplating and similar services pose no great problem, since the load, once properly adjusted, remains static, arc welding is peculiar in that the load is always varying in a random manner.

Consider the very process of striking the arc at the beginning of the weld. The output circuit is directly short circuited. This places a very high current demand upon the rectifier for a short interval. Peak current values of 4 to 6 times normal output rating are not uncommon and under certain circumstances this multiplier may reach values greater than 10.

This condition also exists during the time that the arc is in operation. The resistance of the welding arc is not constant (especially for manual metallic arc with stick electrodes) and varies between wide limits in a random manner within short time intervals. This means that the rectifier in the welder will be subjected to short time overloads at random intervals—a very significant point in the proper choice of rectifiers.

To get sufficient current carrying capacity in our bridge rectifier unit, we must parallel plate area for the desired current density.

It has been found that with forced draft cooling, a current density in the vicinity of 0.6 A/in.^{2}, based on rated welder output, has worked out well in practice. From the design standpoint it is a simple step to determine the number of plates required in parallel.

Continuing with our design example we find that with one-third of the rated current passing through each rectifying leg of the 3-phase bridge, and a current density of 0.6 A/in.^{2}, a total paralleling rectifying area of 166 sq in. is needed. Because of edge masking and mounting arrangements of the rectifier plate, a 10 to 15% reduction of rectifying area must be taken into account. This puts our total area requirement in the vicinity of 190 sq in. If the designer has a 6- x 10-in. cell available, he will choose three plates in parallel.

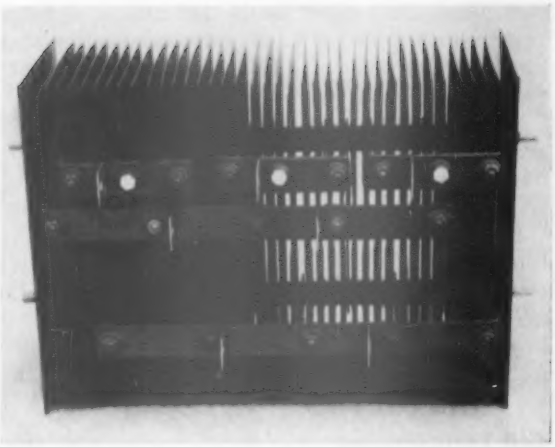


Fig. 1 A typical 300-amp, 3-phase bridge selenium rectifier for welder use. The two long bus bars are the d-c output terminals while the three smaller ones feed in the 3-phase alternating current

Rectifier Arrangements

The whole rectifier stack now turns out to be a 6-element bridge stack with each element consisting of two plates in series for proper voltage, and three plates in parallel for the correct current rating. The stack shown in Fig. 1 is a typical 300-amp welder stack.

Rectifier engineers designate this configuration as a 6-2-3 stack. The first digit identifies the type of connection—in this case a 6-element, 3-phase bridge arrangement. A single-phase bridge is identified by the numeral 4, a single-phase full-wave stack as 2, while a simple half-wave unit carries the number 1. The second digit indicates the number of cells in series while the third digit tells the number of cells in parallel. Incidentally, the product of the three digits gives us the total number of cells in the stack and can serve as a rough index of the comparable cost.

Rectifier Structure

This is neither the time nor the place to go into the detailed functional theory of the metallic rectifier, but it is necessary to understand the structure in order to appreciate the factors determining the life and performance of the selenium rectifier in welder service.

A cross-sectional diagram of a selenium cell is shown in Fig. 2. The back electrode, made of aluminum or

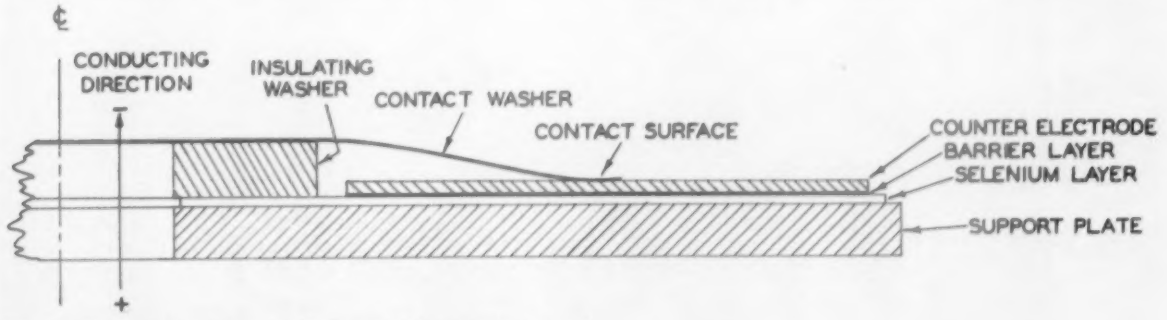


Fig. 2 Cross-sectional diagram (not to scale) of a selenium cell

steel, serves as a support for the thin layer of selenium. The front electrode is a layer of low-temperature alloy whose melting point may lie between 95 and 170° C. depending upon the composition.

None of the materials, in themselves, show signs of asymmetric conduction, but the junction between the front electrode and the selenium, after a "forming" period, does show very decided "one-way" conductivity.

In order to have this "barrier layer," the junction between the selenium and front electrode, function as a rectifying area, very rigid control of the form of the selenium must be maintained. Selenium exists in several allotropic forms and only one, the β crystalline form will work, and the transition temperature points are low and very critical.

Cell Temperatures

It is therefore quite apparent that the temperature of the cells must be kept at a low value. This is accomplished by proper cell spacing and an adequate, streamlined, air flow. Generally an air velocity of around 500 fpm is entirely adequate. Nothing, of course, must interfere with the air flow while the rectifier is in operation if damage is to be avoided.

Keeping a rectifier cool in the over-all macroscopic sense presents no problem but there is more involved. We must look more closely at the minute local heating picture in order to appreciate the significance of keeping all parts of the rectifier cool. This is more or less unique to rectifiers in welding service because of the momentary high current overload surges encountered.

It is generally agreed that rectifier failure occurs solely because of overheating. The idea of voltage breakdown, similar to dielectric failure, has been pretty much discounted except in extremely high voltage situations. What does appear to happen is that at overload voltages the current passed through in the reverse direction (leakage current) causes local high temperature conditions, the greatest heat energy being released in the weakest spot of the cell where the current is highest. If this energy cannot immediately be removed from the local spot by conduction to neighboring areas, failure at that point will occur. Failure will consist of loss of asymmetric conduction, an increase in temperature, a subsequent decrease in the resistance of the selenium and great increase in current flow at the spot and a miniature "explosion." The spot may thus blow clear and result only in a slight decrease of total rectifying area. Typical "blow spots" are shown in Fig. 3. On the other hand, if the failure occurs under the contact mechanism, it is possible that a dead short may occur and complete failure of the cell will be the result.

Fig. 3 Enlarged photo of a section of a selenium rectifier plate showing typical "blow spots"—point failures that have burned free and healed with no damage other than reduction of total rectifying area

Rectifier manufacturers have attacked the problem on several fronts. The first step is to produce a more uniform layer of selenium and thus eliminate the weak spots. Where the nature of the manufacturing process has not allowed this, the weak spots have been deliberately "burned out" under controlled conditions.

Another step has been to increase the contact area between the front electrode and the contact mechanism. This has resulted in a lower current density and reduced localized heating under the contact washer.

The success of the steps is attested by the fact that several of the welder manufacturers place a sliding-scale five-year warranty on their rectifiers. Field tests have shown that if rectifiers are specifically designed for the rugged welding service, the failure record is well within that of other standard electrical components.

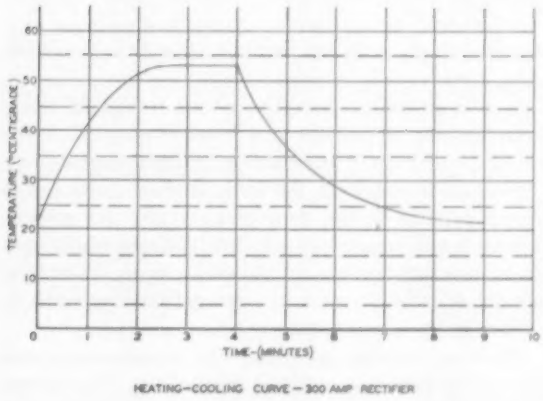
Rectifier Aging

Long-term aging of rectifiers is generally manifested by an increase in forward resistance and a decrease in reverse resistance. The net result is a loss of efficiency.

Whereas sudden breakdown is a result of localized heating, as pointed out before, aging is a consequence of continued over-all high temperature operation. By taking certain steps in the initial construction of the rectifier, the aging characteristics of the stack may be influenced. In many short term applications, aging may be of no significance, but in welding applications it is extremely important and aging must be reduced to a minimum.

Field experience of a specific make of a rectifier type welder over a period of four years has shown only about a 5% decrease in rectifier efficiency.

The rectifiers must be operated within their design temperature ratings. As the operating temperature is raised, the aging process speeds up. It must be borne in mind, however, that aging generally *does not* result in failure directly. Aging is a process that proceeds up to a certain point where complete stability is achieved.



HEATING—COOLING CURVE—300 AMP RECTIFIER

Fig. 4 A typical thermal cycle of a selenium rectifier plate. Rectifier was operated at full power for 4 min and then was shut down. Cooling fan operating continuously

Duty Cycle

The temperature vs. time curve of a selenium rectifier, as shown in Fig. 4, is also of great interest to the welding machine designer. Careful experiments have shown that the steady state temperature of a rectifier is reached in about three minutes (it may vary somewhat for different types of construction). This means that we must rate the rectifier in a different way than we do other electrical components such as transformers that have a high heat capacity.

The reciprocity law between temperature and time, that we so loosely apply in our duty cycle concept, cannot be used indiscriminately without modification in the rectifier case. For example, we cannot effect any rectifier economies in, let us say, a 300-amp d-c welder by reducing the so-called duty cycle from 100 to 60% or to 30%. The rectifier rating must still be 100% duty cycle since the time interval for stabilizing temperature is only 3 min. In other words, overloading a rectifier-type welder by increasing the duty cycle beyond the time rating is, as far as the *rectifier only* is concerned, of no consequence.

On the other hand, overloading a rectifier with higher current than its maximum value for reduced time intervals, unless those intervals are very short (on the order of seconds) is definitely asking for trouble.

Remember now, we are speaking of standard NEMA rated machines and are referring to current overloads beyond the rated maximum output. If a rectifier welder is used beyond maximum current rating at reduced voltage loads, even though the kw output may be down, the rectifier is placed into the rapid aging bracket. This, however, is no different from most electrical devices since the internal losses are roughly a function of the square of the current, and of course the internal losses produce the temperature rise.

Self Capacitance

Another important parameter of the rectifier, that plays an important part in welding, is the self-capacitance. Since the barrier layer is microscopically thin, it is to be expected that the self-capacitance would be extremely high. Of course, the rectifier does not act like a straight capacitor but more like a network of resistance and capacitance in series and in parallel. The fact remains that the capacitance is very high and as a result can exert appreciable influence on the dynamic characteristics of the welding machine. Most workers agree that the average capacitance of a selenium rectifier is on the order of 0.1 mfd per square inch. This may account for some of the observable performance differences between rectifier type welders of varied manufacture.

Efficiency

Another point to consider is the rectifier efficiency. It is impossible to give an absolute figure on rectifier efficiency unless a point of operation is specified.

In any rectifier, the losses are a function solely of the product of the square of the current and the resistance.

The resistance of the selenium cell *decreases* with an increase in temperature so it becomes apparent that the rectifiers should be operated at an optimum temperature to secure highest efficiencies. It should be noted, however, that the selenium rectifier is to some extent self-regulating for as the temperature of the stack decreases the losses increase and the temperature is brought back up.

This feature of the selenium rectifier can be put to advantageous use by the designer of the welding machine in producing a complete unit that will not show the severe characteristic output vs. time drop-off that is sometimes observable in motor-generator sets.

The approximately 5 to 10% efficiency increase for the selenium rectifier observed during an ambient temperature rise of 20° C generally can offset the efficiency decrease of the copper in the rest of the machine. It should, however, be pointed out that beyond this optimum efficiency load there is a slight drop in efficiency with increased loading.

For all general purposes the over-all efficiency of a properly designed 3-phase selenium rectifier stack, operating under optimum conditions, is around 85%. Rectifier designers are looking toward improvements in this direction and some significant laboratory work has been done.

Possible Improvements

Progress in metallic rectifiers can certainly be expected which will improve their electrical efficiency and reduce manufacturing costs. There are many avenues open to the investigator to achieve higher-voltage, reliable selenium cells, so that we can use the ideal 6-1-X configuration for welding power supplies. Progress in this direction will also result in an increase in efficiency.

Cuprous-oxide rectifiers should also be improved in voltage rating but unless this rating can be increased by a factor of 5, the rectifiers would not receive too much consideration for welding service.

If improvements of this type could be achieved with the magnesium-cupric sulfide rectifier and the aging characteristics made better, it might find some applications in welding, especially in view of the fairly high current densities that it can handle.

Nor should we rule out the possibility that some of the many other materials that have been tried, perhaps tellurium, silicon, and compounds such as titanium dioxide, may show very desirable properties.

Germanium

In fact, only recently another type of power rectifier has appeared that seems to show exceptional promise in the field under discussion. It is the germanium power rectifier. Technically it cannot be classed as a metallic rectifier of the same group as those discussed in this paper earlier. The mechanics of asymmetric conduction are somewhat different in the germanium cell than in the selenium rectifier.

Extremely high efficiency figures have been achieved for germanium rectifiers at phenomenal current densities. A cell of 100-amp rating may have an active surface of about the size of a quarter. Voltages are also reported to be much higher than selenium.

This results in what appears to be a very compact, highly efficient rectifier almost ideally suited for certain welding purposes.

The main drawback at the present time seems to be one of high cost and the fact that present production methods are very tricky. As in many very new fields, laboratory progress is being made so rapidly that commercial production techniques are looked at very critically before full commitments are made.

Germanium rectifiers will no doubt find their first commercial acceptance where selenium rectifiers are now used but the need for a much smaller, more efficient and higher voltage rectifier outweighs the higher cost. It would seem that the first extensive use of germanium cells will be in aircraft applications.

In the welding field it appears that the germanium rectifier, because of its high efficiency and compact size, will find its use in high-current, automatic welding equipment.

Germanium will therefore supplement the selenium rectifier rather than replace it. It will extend the range of usefulness of metallic rectifiers by supplying a relatively high voltage—very high current rectifier in a very compact package.

Summary

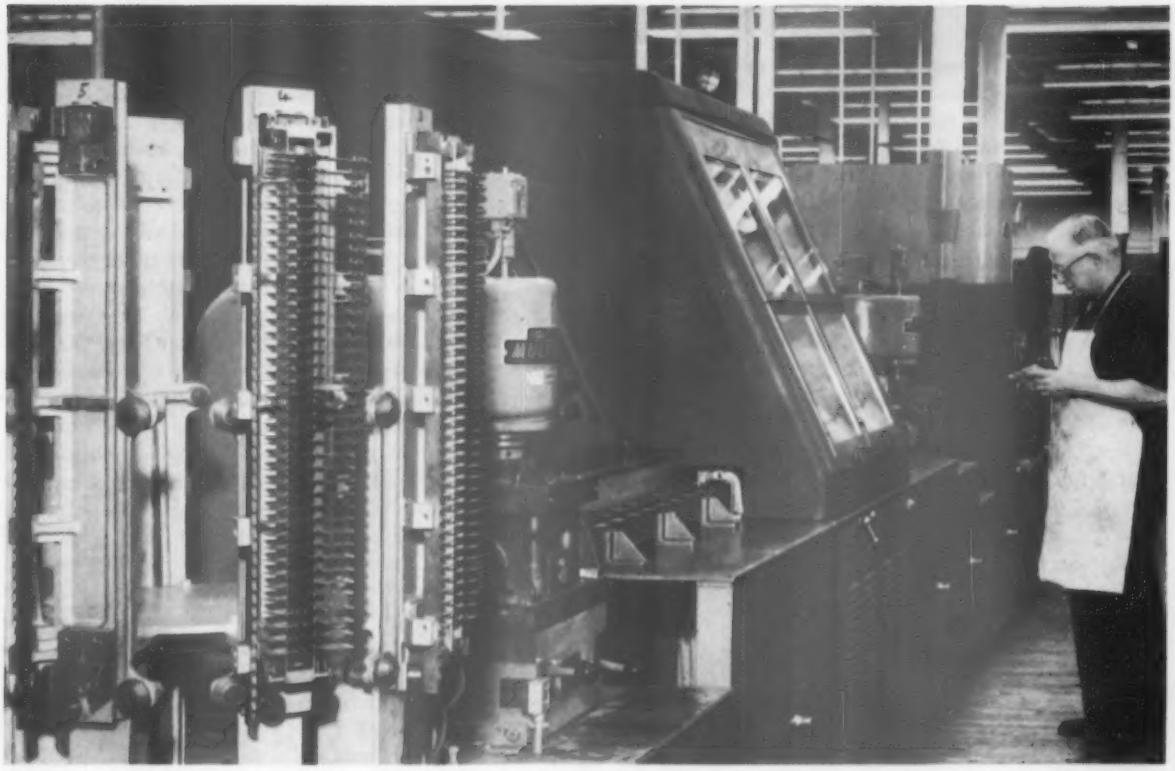
In summarizing then, we can conclude that at the present time the only practical commercial rectifier suitable for arc welding service is the selenium rectifier. It is capable of giving highly reliable, efficient service at a cost figure that places the d-c rectifier type welder with its various advantages into the same price class as the conventional motor-generator set. Also, since the rectifier itself represents the major cost item in a rectifier welder, the prospect that lower rectifier prices would be reflected in welder costs is in the realm of possibility.

The life of a selenium rectifier, if operated within temperature limits, should be on the same order of magnitude as any other static electrical device (such as a transformer, for example). Aging is of no great consequence if the cells are not overloaded temperature-wise.

Improvements in selenium rectifiers should bring about greater efficiencies and lowered costs with the ability to take much more severe abusive overload.

The only rectifier on the horizon that shows initial greater promise is the germanium type. Its present high cost and adaptability to applications which cannot be handled adequately by selenium do ot appear to make it a challenger in the general welding machine market at the present time.

Thus the metallic rectifiers—the copper-oxides and sulfides, the selenium, and now germanium—have grown to the stature of a proved, reliable electrical device—one that can serve industry well, and the welding industry specifically, with distinction.



Operator gaging parts to check adjustment of percussion welder in automatic forming and welding line for relay contacts

AUTOMATIC PERCUSSION WELDING OF TELEPHONE RELAY CONTACTS

Millions of percussion welds have been made during the past year in joining contacts to the ends of the wires for the Bell System's new wire spring relay. Much development effort was necessary to adapt this old but little used method to modern automatic production lines, with many important advantages

BY A. L. QUINLAN

Percussion welding is being used for contact welding on the automatic wire spring relay manufacturing lines at Western Electric Co. This method was chosen for its speed and accuracy of locating contacts. The welder is based on the principles of the half-century-old Vang Process which previously has been used but little in industry. In this process a large capacitor charged to a potential of about 1.5 kv is discharged across the gap between the contact and wire end as the contact approaches the wire at high speed. When the gap has been reduced to a few thousandths of an inch the arc initiates itself and the energy is expended in a fraction of a millisecond. Metal in the gap to a depth of 0.005 to 0.010 in. has been heated and expelled so the continuing movement brings the hot

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Presented at the Thirty-Fifth National Fall Meeting AWS, held in Chicago, Nov. 1-5, 1954.

surfaces together with percussive force and the weld is made.

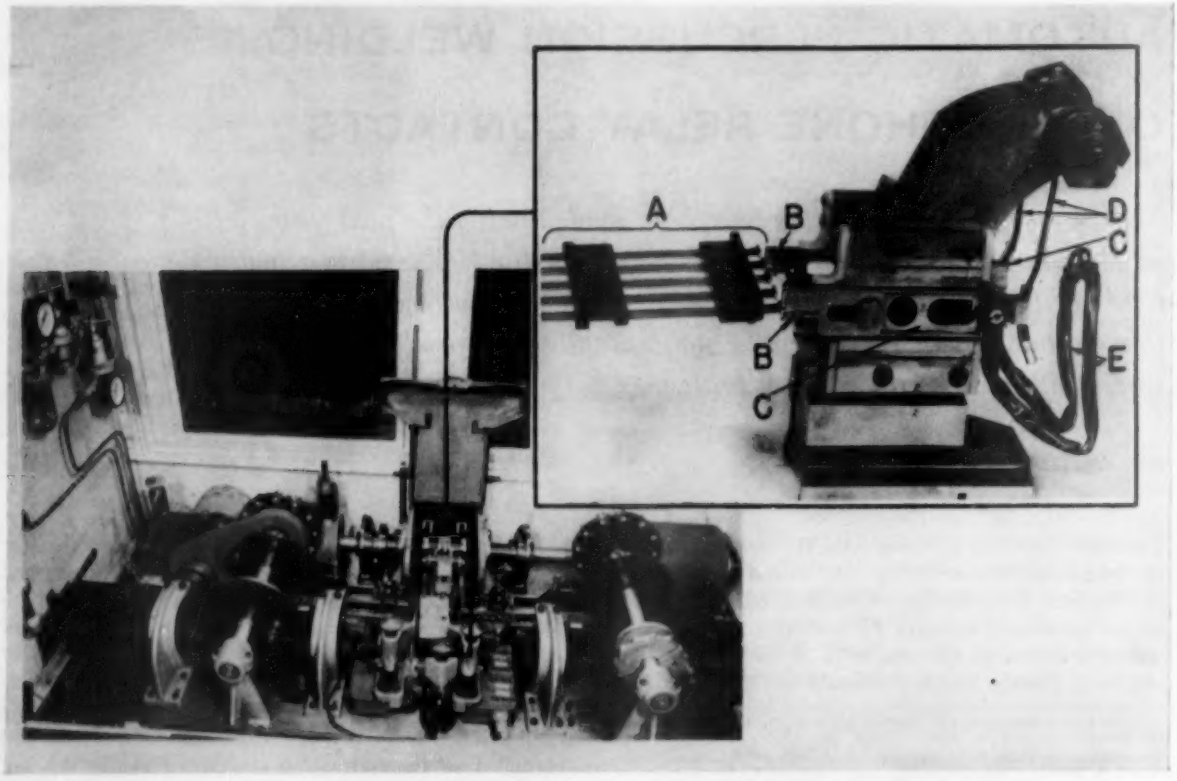
Previous articles by the author have dealt with early development work and the establishment of necessary conditions for good welds. This paper deals with advanced work on later problems and how they have been solved by improved welder design. The relay part to which the contact blocks are welded is called a "comb" as shown in the enlarged portions of Fig. 1. The contact blocks are welded to the ends of the 15 wires extending in a parallel array about one-quarter of an inch from the plastic. The palladium contacting surfaces of all contacts must be in the same plane within ±0.002 in. to meet the design requirements. The wires must be well anchored in the plastic to resist the impact of welding.

The improved features in the gun structure, flash protection and faster capacitor charging contribute to the quality of welds and the life of the welder. Although production figures on this design are limited to a few months' operation, good weld strength is indicated by the chart covering 150,000 welds.

Welder Design Features

The heart of the welder is the double-gun mechanism on which jaws are mounted for holding contacts so two can be welded at a time, Fig. 1. Each gun is a separate member, providing accurate movement. A contact is inserted in the jaws by a transfer finger near the back of the gun stroke and transported to the wire end where the weld is made. The jaws are simple in design and easily replaceable. The contact blocks are cut out of bimetal tapes by notched punches located on each side of the gun units. The tape consists of a palladium cap roll welded to a cupro-nickel base in a previous operation. It enters the welder from reels on either side. The punch notch neatly fits the tape to prevent distorting it when the cut is made and sides of the notch hold the contact as it is carried to the end of the stroke where a transfer finger pushes it in a lateral direction out of the punch and into the jaws. A slight bumping motion of the gun against the transfer finger seats the contact against the back of the jaws. When a gun is released by its cam on the main shaft it springs forward and strikes the contact against the wire end. The 30-g gun is driven by a spring exerting about 5 lb of force to propel it to a velocity at impact of approximately 30 ips. Each gun is connected to its cam by an intermediate sliding link. Sufficient time is provided before the cam returns the gun thus pulling the jaws off the welded contact. This operation puts a 2-lb test on every weld. Because of the arrangement of the wires in five groups of three the welding cannot be divided equally between two guns. An odd welding sequence is used with the welding progressing in pairs to finish the comb. The comb-supporting table is stepped and accurately positioned for each weld or pair of welds by a cam and lever system.

Fig. 1 Top view of percussion welder (*left*). Enlarged view of guns and wire comb in weld position (*above*). *A*—Wire comb; *B, B*—jaws; *C, C*—guns; *D*—gun springs; *E*—flexible leads



One cycle of the main cam shaft at the back of the machine welds one contact with each gun except for the first three cycles when the No. 1 gun is idle. Each cycle requires $^1/_2$ sec and 9 cycles are required for 15 contacts with 3 more cycles for unloading and reloading the nest, making a total of 12 cycles or 6 sec per comb. Each cycle includes the feeding of the tape, the punching of a part out of it , the transfer of the part to the jaws, the advance of the jaws to the "drop-off" point, the release of the gun, its travel to the wire end, a slight dwell for the metal to freeze, then the return of the gun to the start of the stroke.

The lower cabinets at either side of the welder proper, see Fig. 2, provide space for tape reel supports just under the table tops, tape reel storage below and to the rear, electrical control relays and electronic components at either side. Under the welder table top is the 3-hp motor, drive gears and the capacitors. Compressed air is provided for clearing the weld zone of arc products which might affect the initiation of subsequent arcs. Compressed air carrying an oil mist is introduced into the gun bearings through drilled holes in the supporting structure to prevent the entrance of flash which might affect the speed of the guns. Also the gun mechanism is covered and an exhaust system is provided to keep the arc products out of the weld zone and the bearings. The guns do not weld at exactly the same time. The No. 1 gun is released milliseconds before the No. 2 gun so there will be no mechanical or electrical interference between them.

Fig. 2 Percussion welder for multi-contact relays with doors open to show electrical equipment

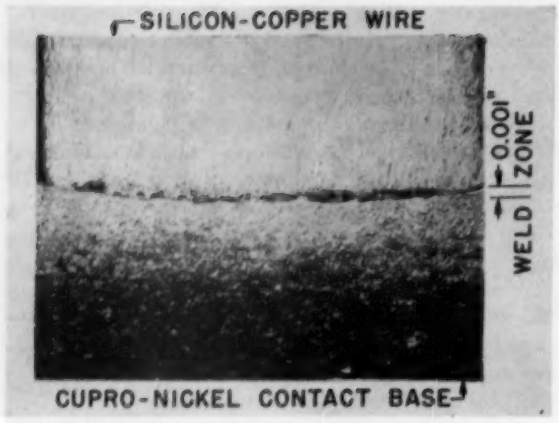


Fig. 3 Micrograph of typical weld section

Electrical System

The welding is done by two 75 μf capacitors in each discharge circuit to a gun. The resistance and inductance values of the circuits are adjusted to produce the best weld pulse. The current is controlled in this way to produce a peak value for the required duration under conditions having been found to be optimum for good welds. High peak currents tend to burn contacting surfaces. Also, burning may be caused by a longer duration of current flow which extends the period to the point where it is occurring when the welded contact is shifting in the jaws because of vibration. Multileaf springs are used for completing the weld circuits to the wires being welded. Contact is made only during the welding interval after the nest has been positioned at the wire centers. The capacitors are charged through thyratrons by a controlled rectifier voltage after the jaws have been retracted from the previous welds.

Safety

Operator safety has been provided by enclosing all cabinets which give access to the high voltage, including the working top of the machine. The enclosing doors are equipped with door switches so the high voltage cannot be turned on with the doors open. Also the magnetically retracted shorting mechanism on the capacitors is released when this circuit is opened. In addition, a manually operated handle mechanism for unlocking the doors forces the safety shorting mechanism onto the capacitor terminals when the doors are unlocked.

Safety for the machine is provided by a series of microswitches controlling the seating of the part in the nest and the feeding of short contacts which might burn up in the arc and injure the jaws. Normally the end of the wire of the comb will not contact the back of the jaws when the gun is released so the arc is not initiated when no contact is in the jaws. Failure to seat the part in the nest might cause a wire to contact the jaws and burn them. If either of the above conditions occur, the test which is made by a cam at the proper time in the cycle detects the fault and the safety thyratron discharges the capacitors before the parts come together for welding. Should the piece part transfer conveyor be stalled from any cause the drive pulls free at a preset load and a microswitch stops the machine.

Weld Quality

High-speed percussion welding heats little metal and leaves only one or two thousandth of an inch thickness of metal in the weld zone that shows any effects of having been heated, see Fig. 3. Thermal or electrical conductivity of the metal has little effect on the weld because of the shortness of the time involved. A typical weld strength distribution is shown in Fig. 4. These results are obtained from a gage which pulls on one side of the contact about 0.010 in. from the side of the wire, thus subjecting the weld zone to a combined tension and bending stress. A limited comparative test in straight tension showed more samples to fail in the wire outside of the weld zone at 150 to 160 lb than in the weld. Gas pockets in a critical part of the weld are probably the greatest cause of lower weld strengths.

Conclusions

After making millions of welds on the wire spring type of relays, percussion welding has been proved to be well suited to the job. Accuracy of location and good weld strength are being obtained. This method is especially useful for high-speed automatic welding and where joints must be made between dissimilar metal. Metals of high thermal conductivity and high electrical conductivity join readily.

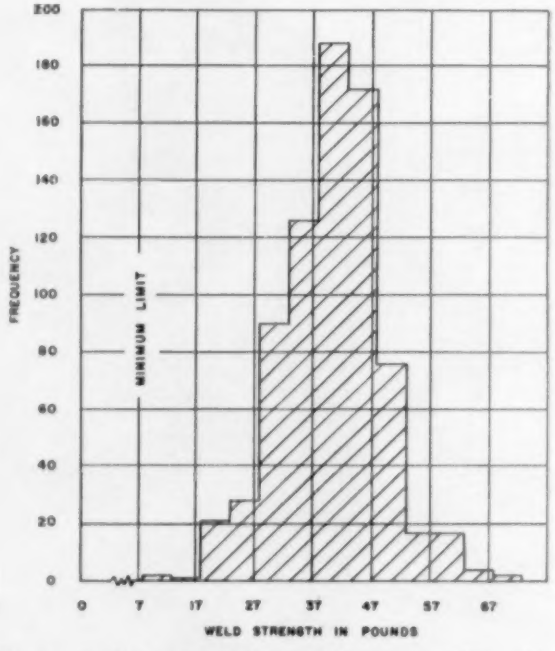
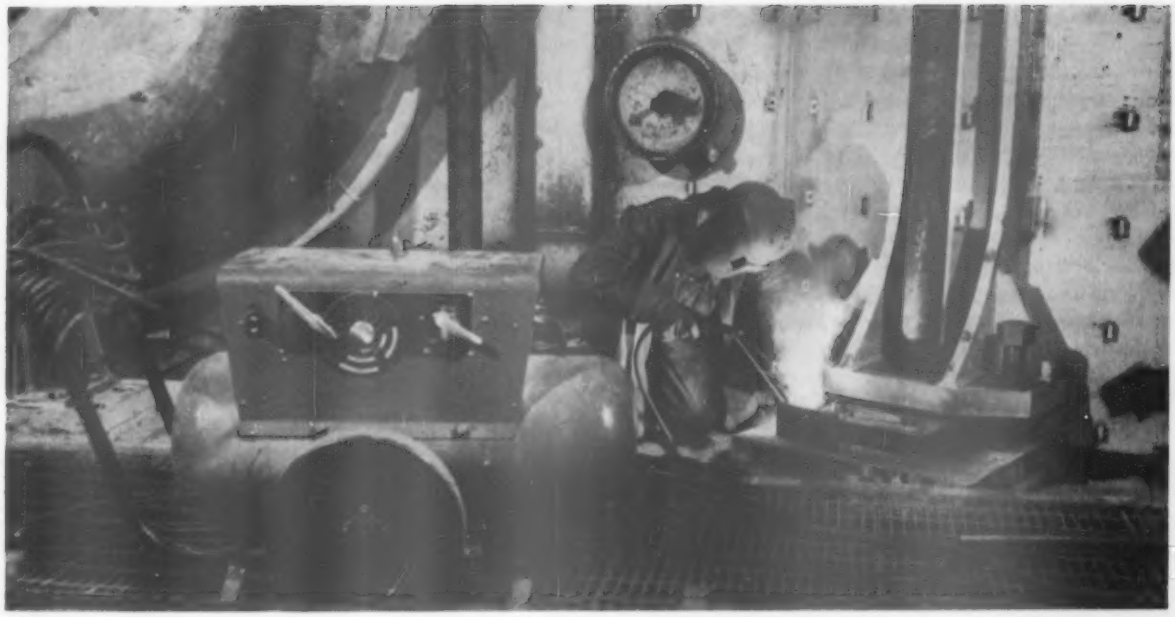


Fig. 4 Typical weld strength distribution for 738 welds tested from a production of 150,000 welds



Versatile d-c welding is widely used for fabrication and installation of large equipment

SELECTION OF D-C ARC WELDING POWER SOURCES

Some of the lesser known factors involved in choice of rectifiers versus motor-generators d-c arc welders are presented

BY A. U. WELCH

The present availability of two quite different types of power sources for conventional metal-arc direct-current welding indicates the desirability of a careful study of the relative merits of motor generator type versus rectifier type arc welders. There are indications that many users do not have complete information on the factors affecting a choice of welder, and the purpose of this paper is to discuss some of the factors on which there has not been much information previously presented. Electrical efficiency differences are shown to be of insufficient economic importance as compared to usability to control choice. Slight differences in usability affecting labor cost are of more importance.

Electrical transient performance of the two power sources is discussed and it is shown that transient currents and voltages have a controlling influence on usability of the welder in hand metal arc welding. High quality short arc welding and position welding require substantial current peaks for dissipating short circuiting metal, and transient current affects ease of starting. Voltage transients determine arc stability and continuity. Either rectifier welders or generator type can be designed for good welding performance in these respects and some broad aspects of designs are discussed.

Factors Affecting Choice

In order of importance to the user of a d-c welder the factors affecting his choice of a generator type or a rectifier type are usability, dependability and electrical efficiency. His major welding cost is the operator's wages and this factor can be affected substantially by certain not generally understood usability attributes of

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the power source. Dependability is of obvious economic importance in its affect on maintenance cost and losses from downtime. Electric power cost is only a few percent of total costs, and differences between the two types of welder are insignificant compared to possible usability differences.

Efficiency

The full load losses in a 400-amp rectifier welder are on the order of 8 kw and approximately 1 kw higher in a modern motor-generator welder, efficiency therefore is in the range of 64 to 67%. At no load, a motor-generator may consume 3 kw more than the rectifier welder. This is a more important difference, especially so since most welders idle more than 50% of the time. A three- to five-cents an hour power cost difference can be offset several times if there are substantial usability differences affecting the quantity or quality of the operator's output.

Dependability

Dependability of the two types of equipment is determined largely by the design practices of the manufacturer. Type of rectifier used, current and voltage loading of the plates are critical factors; ventilation of rectifiers is another important point.

The time constant of dry plate selenium rectifiers is very much shorter than that of a direct-current generator, particularly at high overloads. At maximum output, rectifier stacks may reach ultimate constant temperature in a matter of 2 or 3 min and the corresponding time interval for generator and commutating system may be 15 min. This is the reason that many customers have chosen to purchase 400-amp rectifier welders to do the jobs they were accustomed to doing with a 300-amp motor generator. For the applications where many short-time jobs are done at more than 100% of rated current, the motor-generator welder offers substantially lower investment because of the smaller rating required.

Selenium rectifier stacks used in welders deteriorate with service. Aging increases forward drop and heating of a cell and, although reduction of output can be compensated far more readily in the welder than in other types of rectifier applications by simply moving the current adjuster, there is still a limit of life of approximately 10,000 to 15,000 hr of rated load service beyond which the plate losses cause such high operating temperatures as to result in failure. The owner of a rectifier welder, therefore, can look forward to the necessity of complete replacement of rectifier stacks in a number of years, the time being determined by a combination of the conservativeness of the manufacturer and the type of service to which the welder has been subjected.

Dry-type rectifiers are particularly vulnerable to overvoltages, either steady state caused by high power line voltage or transient overvoltages which occur in any electrical apparatus subject to fluctuating loads. Switching in the d-c welding circuit is particularly likely to cause high transient voltages. Even though users of rectifier welders are warned to avoid putting switches in the d-c circuit, substantial inductive kick voltages are caused by scratching electrode to start the welding arc, especially in the current range from 100 to 200 amp. The first contact of a moderately large cold electrode to a cold plate is usually of such short duration as not to bring points on the two pieces of metal to a temperature which will emit electrons or conductive gases. The first spark, therefore, is quickly extinguished, so quickly as to cause large induced voltage by the rapid rate of change of current flowing through the inductance of the welder. Some types of selenium plates are partly self-protecting in that the weakest spot in a plate sparks through, the spark burning away the counter electrode to the point where the spark is extinguished and the plate has suffered only a small loss of effective area for carrying current. The spark-through, though, has removed the moisture protection of the heavy varnish film that the manufacturer has carefully applied to protect the cell. Other types of plates are made in such fashion as to be self-clearing from internal sparking without breaking through the protective film. A spark-through that happens to occur under the current collecting washer usually results in sudden death to the rectifier. Another approach used with great success has been to manufacture a plate with a substantial degree of what might be called saturation, in other words the plate inverse volt-ampere curve bends over very steeply at a voltage perhaps 10% above operating voltage so that excessive voltage results in high uniformly distributed leakage current throughout the plate. The order of magnitude of this current can be made high enough to absorb the inductive kick energy without allowing the voltage to rise high enough to cause rectifier damage. There are other technical possibilities for protecting rectifiers from inductive surge voltages. Unfortunately they are rarely economically practicable.

Of the three requirements of a good welder, usability

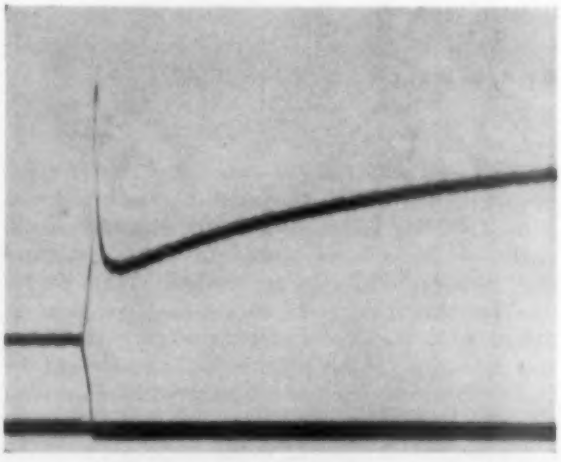

Fig. 1 Oscillogram of motor generator arc welder at interruption of current by switch

—and consequently what might be called weldability—is the most important. The largest expense in metal arc welding is the operator's pay check and, compared to this, amortized investment and electric power consumption are minor, almost negligible items of operating costs.

Usability

Welding current output of the proposed new NEMA design standards parallel approximately the motor-generator standards which have been used for many years by practically all manufacturers. Some rectifier welders now on the market in the industrial ratings do not meet the present motor-generator standard of producing 125% rated current at 40-v load, having maximum output of only 100% rated current at that point, and consequently are limited in application and cannot be considered as a direct substitute for a motor-generator welder.

Arc Blow

Magnetic arc blow is a factor not well understood in d-c arc welding. It is the major limitation on size of electrode and current usable in many applications, and therefore limits speed and controls cost. It would be pleasant indeed to be able to report that the newer rectifier welders offer an improvement in this respect over the old familiar welding generator. Unfortunately tests do not substantiate this, and a little reflection indicates why. Arc blow in welding steel is caused largely by either temporary or permanent magnetization of the work by the welding current flowing through it. A-c welding greatly reduces arc blow because ac does not permanently magnetize steel, and ac temporarily magnetizes steel plate less than dc because of the demagnetizing effects of induced alternating current in the plate. In other words, the plate can carry a-c magnetic flux in only a thin film on the surface, and not throughout its thickness. DC is dc whether produced by a rectifier or a generator (after all, the generator generates ac and rectifies by means of its commutator). DC magnetizes steel sufficiently to produce motor action on the welding arc stream regardless of its source. A superposed a-c ripple does not change the net force on the arc caused by the direct component of current and the undirectional magnetic field. From the usability standpoint of arc blow the two power sources are identical. Static volt ampere curves of the two types of welder indicate no differences of enough magnitude to cause measurable usability differences, with one exception to be noted later. There are or can be large differences in dynamic performance which for conventional stick metal electrode welding can cause great differences in usability. Dynamic or transient *voltages* can have substantial effect on arc stability.

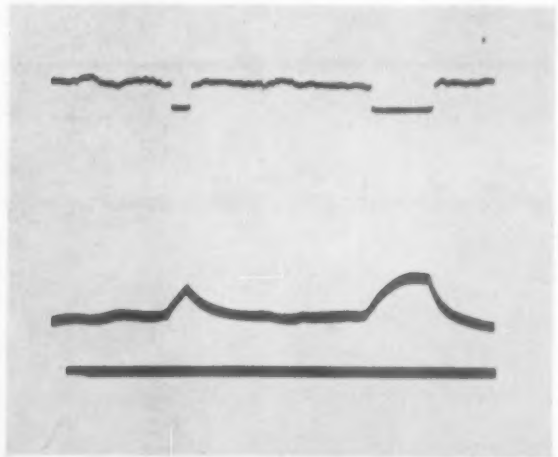

Fig. 2 Oscillogram of voltage and current of arc welder showing intervals of metal transfer. Upper trace arc voltage lower current

Voltage Transients

Electrical arc stability—lack of sudden current stoppage—is about equal in well-designed welders of the rectifier and generator type. The recovery to open circuit voltage is very fast in a rectifier welder (limited by rectifier and lead capacitance) after a momentary outage. The inductive overshoot assists in restarting an unstable arc. The plate capacitance and leakage or sparking limit the magnitude of overshoot and may affect arc stability in special applications. On the other hand, the inductive kick on a generator welder does not have this type of limitation but arc stability may be adversly affected by the sag of voltage after the inductive kick (Figure 1).

The oscillogram shows current cut off by a switch; following this an inductive transient overvoltage, a sag to the then generated voltage, this latter slowly rising to open circuit voltage.

Tests have indicated that popouts generally occur immediately after one of the momentary short circuits which occur while metal is dripping from the electrode to the work. During the short circuit the metal is

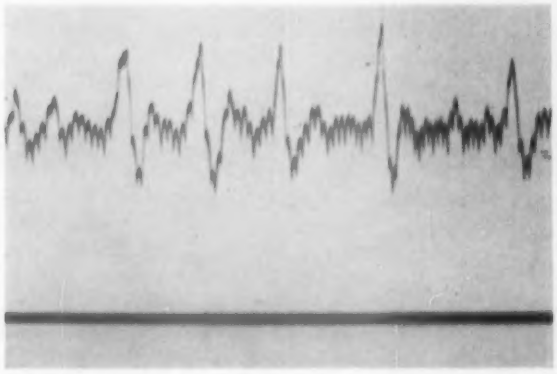

Fig. 3 Oscillogram of rectifier welder during welding, current trace shown soft arc type

cooling, a large drop of metal falling will leave a long comparatively cool gap difficult to ionize without higher than normal arc voltage. The power source must deliver enough voltage to cause enough current flow through the cool gasses to reverse the cooling tendency or the arc goes out. A well-designed generator does this by providing sufficient inductive voltage to steady the arc at its weakest time, continuing long enough to allow the generator voltage to build up to higher than normal arc voltage to sustain and reheat the arc. A large number of arc stability tests have shown very good correlation of arc stability with minimum voltage after interrupting a short circuit, providing a substantial inductive surge voltage is also available immediately after the short circuit. The tests also show that a sag to or below normal arc voltage will cause outages regardless of inductive kick.

Dynamic or transient currents greatly affect usability in a very different way.

Current Transients

If all welding arcs were continuous arcs such as tungsten welding arcs, transient currents would be of little interest except during the interval of short circuit at the start. Examination of oscillograms and a high-speed motion picture of conventional metal arc welding at any usable arc length shows frequently repeated short circuits caused by metal dripping across the arc. Welding tests show that the transient currents caused by these short circuits affect weldability very noticeably, particularly the ability to hold short arcs without sticking.

If the momentary current at short circuit is only slightly higher than that during arcing, the welder generally refers to the power source as a "soft arc" machine. Conversely, a machine which delivers current which builds up rapidly to a high peak during the metal transfer the arc is referred to as forceful or "peppy." In this case the operator can crowd or shorten the arc without sticking, a very important consideration on welding heavy metal, welding vertically or in welding with low-hydrogen or similar electrodes where short arc is required for high quality results. The high transient current, slightly by thermal effect, but mostly by pinch effect and other magnetic forces, removes the molten metal bridge quickly before it can chill and freeze the electrode to the work.

Incidentally such a power source provides a few percent higher metal deposition rate by shortening the intervals of short circuit. Heat input is negligibly small during short circuit; power is the product of current and voltage, the current is high but the voltage is nearly zero. A greater percentage of true arc time, therefore, increases welding speed.

It has been a characteristic of most arc-welding generators to furnish a substantial peak current during metal transfer. Figure 2 indicates a ratio of 1.8 times welding current. Some of this characteristic is inherent in the design of a generator in that induced currents in the generator fields tend to resist rapid change of generator voltage; therefore, the generator temporarily puts out more than its steady-state short-circuit current. Early rectifier-type arc welders were of the soft arc type; that is, low short-circuit current as illustrated by Fig. 3. The next step in improvement of rectifier welders was to incorporate auxiliary devices to increase either the steady state or transient short-circuit current. One method is to use paralleled rectifiers supplied from low voltage sources (less than normal arc voltage). This provides a voltage ampere curve with a sudden change of direction below normal arc voltage toward a flatter curve. The transient

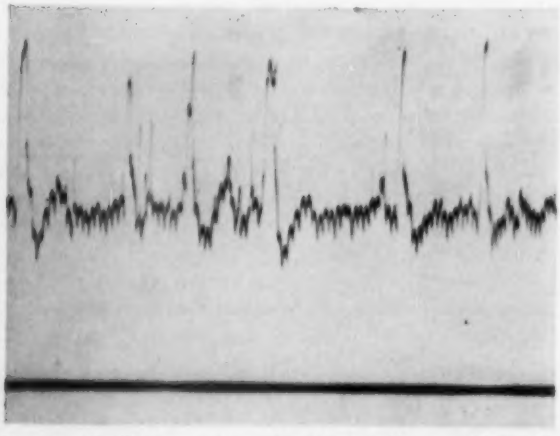

Fig. 4 Oscillogram of current of rectifier welder with increased arc force

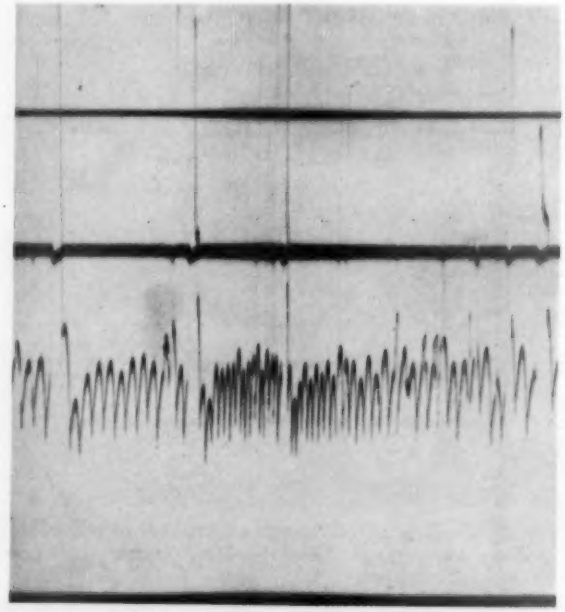

Fig. 5 Oscillogram of rectifier arc welder current with high short circuit transient current. (Two upper horizontal lines not involved in current)

currents delivered by such a welder during the metal transfer is indicated by Fig. 4. It will be seen that the ratio of short circuit to normal current has been increased. The steady welding current is approximately 100 amp. In Fig. 3 the peak short-circuit current was about 150 amp and in Fig. 4 the short-circuit current was about 200 amp. Another approach toward providing high transient short-circuit currents is illustrated in Fig. 5. In this case the peak short-circuit currents are in the order of 300 amp or more. This result is obtained without the problems entailed by continuously operating rectifiers by the approach of using a capacitator discharge circuit. A bank of high capacitance low-voltage d-c capacitators charged by a separate circuit to a value slightly below arc voltage is connected to the welding circuit through a blocking rectifier so that the capacitator bank can discharge through the weld whenever a transient short circuit occurs. Since the transients are intermittent the energy required for dispersing the bridging metal can be stored during the normal arcing periods at a comparatively low power level. This level can be kept low enough so that the circuit furnishing the power can be continuously rated and the circuit, therefore, does not require any protective devices to prevent damage during a continuous short circuit. By means of this circuit it is practicable to obtain a more forceful arc than is economically practicable by a straight parallel rectifier.

The results can be dramatically illustrated by welding tests with the arc force device operating. The user is asked to shorten the arc as much as possible and is able to weld with a very short penetrating type arc. If, while he is welding, the arc force device is disconnected, the operator instantly finds the electrode frozen to the work, indicating that he would have had to hold a much longer arc in order to operate without the arc force control. Avoidance of sticking in metal arc welding leads to definitely improved weld quality. Shorter arcs reduce hydrogen pickup from moisture in the air and there is absence of faults caused by frequent stopping and restarting.

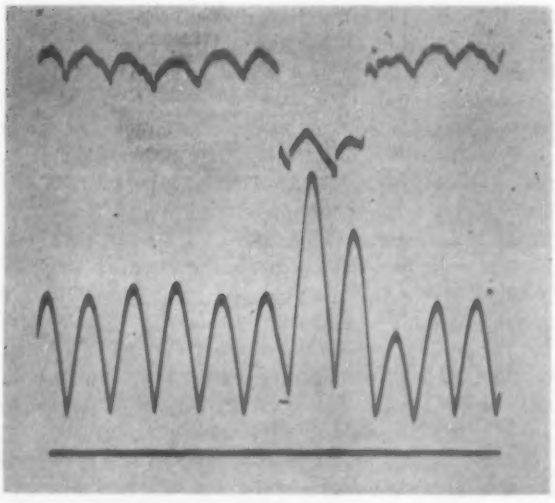

Fig. 6 Oscillogram of rectifier arc welder with saturable reactor control and feed back

The arc forcefulness of any type of d-c welder can be materially reduced if long welding leads are used in the circuit. The welding lead impedance can become such a large part of the total circuit impedance as to smooth down or reduce the transient currents to the point at which the welder becomes a soft arc type machine. This approaches the conditions obtained with the old fashioned constant potential welding set up with resistor boxes for control of current. Where a forceful arc is required the welding machine should be located as close as practicable to the work.

A rectifier welder using saturable core reactors for welding current control can be made to have varying degrees of arc forcefulness, in fact it can be a problem to avoid such violent transients as to cause excess weld spatter. However, some presently available commercial welders of this type do not indicate the presence of sufficient transient current to provide adequate arc force. Figure 6 is an oscillogram of a type which has some part of the feedback circuit so arranged as to couple a substantial third harmonic into the welding current. The lower trace is current; the upper, arc voltage. It will be seen that transient current is only about twice the welding current which would classify the welder in about medium range of arc forcefulness and it will be noted that the duration of short circuit on this oscillogram is comparatively long, indicating a soft type of arc. Welding tests have not indicated any advantage in usability caused by the unusually large current ripple. In fact the minimum current is so low, about one-third of the average current, as to approach the limit of arc stability in critical applications. Arc stability here starts to approach that obtained by an a-c welder rather than of that expected from a d-c welder.

Conclusions

1. This paper has presented some of the lesser-known factors involved in choice of rectifier versus motor-generator d-c arc welders.

2. Electrical efficiency is a minor factor compared to usability characteristics.

3. The purchaser of the equipment should investigate carefully the comparative welding current range and suitability for high current service.

4. Maintenance and replacement cost of parts are well known for motor-generator welders; rectifier welders are too new for complete evaluation, rectifier stack life is the largest undetermined element in maintenance cost.

5. The soft arc characteristic of the simple rectifier welder is entirely adequate for such applications as tungsten-arc gas-shielded welding and sheet metal welding. The user fabricating heavy steel weldments should investigate very carefully the relative arc forcefulness of motor-generator and rectifier welders for his application.

FLASH WELDING HIGH-STRENGTH ALLOY STEELS

DISCUSSION BY J. J. RILEY AND J. H. COOPER

Since Mr. Fassnacht is one of a small group of highly experienced (over 15 years) persons in this particular field, it is most difficult to discuss this paper in any other manner except a highly complimentary one.

Realizing the difficulty of presenting every minor point in a paper on such a broad subject, the following points were probably omitted by Mr. Fassnacht in the interest of brevity, but are offered here in the hope they may be of assistance to someone using this process.

1. (*a*) When using a protective atmosphere, sufficient time of gas flow prior to flashing and upsetting should be provided to thoroughly purge the interior of the welded assembly. (*b*) The gas should be dry, even if necessary to flow through a drier, and the presence of moisture should be particularly checked when using city gas. (*c*) The use of a simple flow meter will assist in maintaining uniformity of protective atmospheric conditions.

2. (*a*) In reference to the tensile tests of coupons reduced in cross section at the weld area and heat treated to strengths of parent metal in excess of 200,000 psi, care should be taken to round corners and provide smooth surfaces in the vicinity of the weld. Lack of such care will produce erratic test results, because these steels are sensitive to such surface imperfections when treated to these strength levels. (*b*) Provided that the tensile coupons show fractures through the weld line at strength levels and uniformity of comparable tests made in the parent metal, there is no reason that the weld should be considered either *better* or *poorer* than the parent metal, and fractures can be expected to take place either in the weld line or external to it, *completely at random*, and so be considered completely acceptable in either case. It is interesting to note in passing that no other welding process can even approximate this quality under similar conditions.

3. Mr. Fassnacht mentioned the use of antispatter compound inside of tubular work. This can be supplemented by snapping disks of heavy asbestos paper in the end of each part a few inches either side of the weld and thus confining the flash to a narrow zone on each side of the center line of the weld.

4. The writers agree with Mr. Fassnacht's probable causes of penetrators. A word of caution should be introduced when translating data compiled on the basis of wall thickness from machine to machine. If it is assumed that the magnitude of energy level is a cause, it is necessary that the weld schedule be readjusted dependent on the size of the equipment. It is impossible in the larger welders (due to geometric clearances) to produce as many amperes (welding current) per secondary volt as in the smaller sizes. Thus the amperes per square inch (based on rated welding capacity) available in the larger size at the same secondary voltage is inherently less and to maintain a satisfactory condition of flashing generally means operating at a slightly higher level of secondary voltage than that possible on a smaller piece of equipment. This in turn means that the amount of burnoff should be increased (and a corresponding increase in flashing time) so that the temperature gradient is not too steep. The purpose of decreasing the steepness of the temperature gradient is to compensate for the larger voids that may otherwise result. Briefly, the above amounts to increasing the amount of burnoff, for the same wall thickness, as the equipment becomes larger in order to compensate for the difference in machine characteristics.

It is normally advisable to use greater burnoff and flashing time than the possible working minimum. From experimental work at RPI, using parabolic flashing cam rise, there was definite evidence that the work attains a definite temperature gradient after a specific amount of burnoff, but further burnoff did not cause any measurable harm. Therefore, under production conditions in which anyone of a number of variables may go slightly astray, it is advisable to allow this extra burnoff as a safety factor.

In conclusion, we feel Mr. Fassnacht has presented a paper of highest merit and one which makes a substantial contribution to the available knowledge of this subject.

J. J. Riley and J. H. Cooper are associated with the Taylor-Winfield Corp., Warren, Ohio.

Paper by W. G. Fassnacht was published in the October 1954 issue of The Welding Journal, pp. 937–944.

AUTHOR'S REPLY

The comments submitted by J. J. Riley and J. H. Cooper are very much appreciated. While this author has no objection to any of the points, the following comments on comments might be in order.

1. When using a protective atmosphere, it seemed too obvious that unless the gas reaches the faying surfaces, no good can be accomplished. So did the use of a flow meter. How else could 10 cfh per inch of diameter be measured? Regarding drying the gas, it seems like a good idea, although we have no data on it. Our gas supply has been reported to have a dew point of −20° which makes it quite dry. We have occassionally made welds before the antispatter compound dried completely and no known difficulty has ensued. However, no specific attempt was ever made to find out if it was deleterious, so it is possible that there could have been a somewhat undesirable effect that escaped notice.

2. The method described for making test coupons is a compromise on the most desirable design for a tensile sample because the cost of producing such is so high. We have not taken special pains to round corners (although it may be necessary as tensile strengths go higher) because we have felt that the influence of corners has been more on elongation rather than on indicated strength and the strength has been our criterion for acceptability.

3. Care should be used when using disks for confining internal flash. If they are put too close together, the flash will build up and could cause a freeze.

4. The author is very happy to have this comment. It is stated in the paper that seldom can cycles be transferred directly from machine to machine, and Messrs. Riley and Cooper have given some specific reasons why.

NEW CONCEPTS IN SPOT X-RAY OF WELDED STRUCTURES

DISCUSSION by A. B. WILDER

The author has stated radiography's greatest value lies in the control of weld quality and then has described a practical means for spot radiographic inspection of welds. In the manufacture of large diameter welded pipe for gas lines we have employed spot radiography. This technique has been found to be practical in the control of weld quality in the longitudinal automatic submerged-arc-welded seam. The technique is also readily adaptable to the control of weld quality in girth welds which is used by the manufacturer for joining two lengths of pipe. In the girth welding process automatic submerged arc welding is also employed.

Destructive mechanical tests are commonly used in the manufacture of welded pipe for the evaluation of weld soundness. The use of spot radiography eliminates the unnecessary expense of conducting these tests. It also provides a convenient means of correcting manufacturing procedures before unsatisfactory welds are produced.

The proper interpretation of radiographs is, as the author has stated, "extremely important." It is necessary, therefore, to provide suitable standards which recognize the fact that in a welded product discontinuities occur which are not necessarily harmful, although they may be undesirable. Practical standards have been developed for welded line pipe. These standards are being used as a manufacturing control in spot radiography with the result that weld quality has been improved.

A. B. Wilder is Chief Metallurgist, National Tube Division, U. S. Steel Corp., 525 William Penn Place, Pittsburgh 30, Pa.

Paper by Harold Hovland was published in THE WELDING JOURNAL, 33 (10), 962-965 (1954).

AUTHOR'S REPLY

Mr. Wilder's comments are indeed appreciated. Utilizing the radiographic tests in proper sequence with the fabricating process will result in lower manufacturing costs and assure a more uniform product.

WINTERTIME IS WELDING TIME FOR THE FARMER

BY M. MISKO

Harvesting is almost over for many farmers. When the crop is in, the farmer has a chance to look over the damage that months of hard work have done to his equipment. Unless winter crops are planted, farm machinery will not be used again until spring planting. Making repairs during the winter will not interfere with farm routine.

Much of the normal wear and tear on equipment can be remedied by oxy-acetylene welding and cutting. Broken hitches, bent frame members, worn gear teeth, cracked axles and broken wheel spokes are but a few of the items that can be repaired with oxy-acetylene equipment. By repairing, rather than replacing parts, the farmer saves time as well as money. In many cases repairs are made without dismantling machinery.

The accompanying pictures show a few repairs that were made on a farm, when the equipment was not

M. Misko is associated with Linde Air Products Co., Pittsburgh, Pa.

Fig. 1 Braze welding was used to repair this tractor hitch when it broke. The corn picker was later used to harvest over 100 acres without the least sign of fracture or cracking of the weld

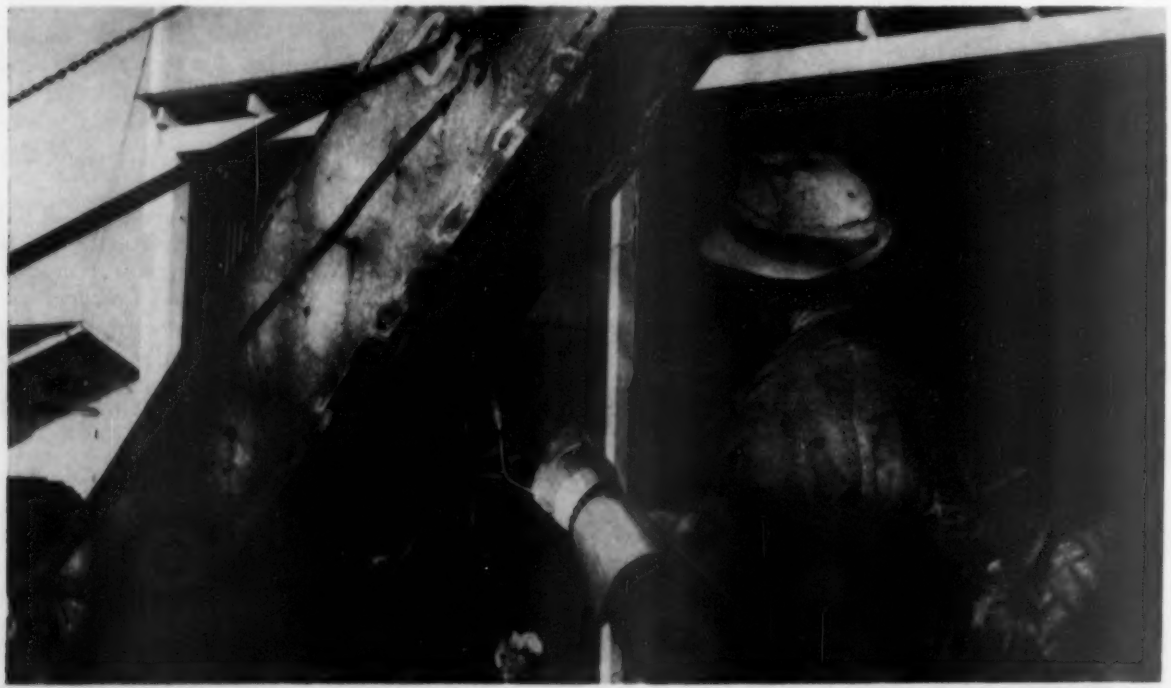

Fig. 2 This corn-picker elevator was damaged when it bumped against the wagon box. The elevator was straightened by heating and 1-in. angles were welded to the bottom

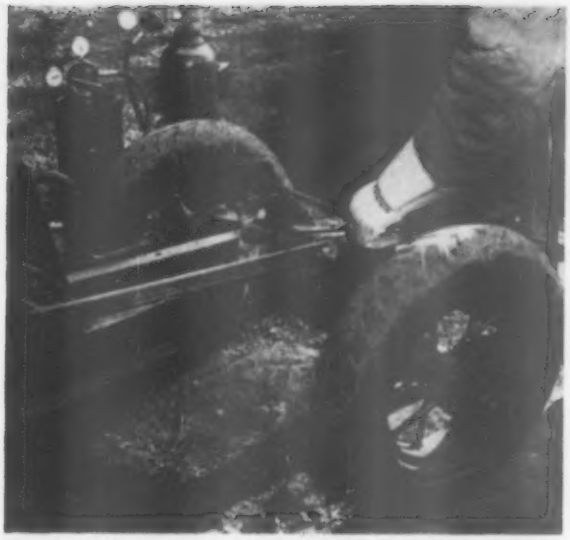

Fig. 3 When this cutter-bar adjustment-plate broke, the weld was made without removing the plate from the combine. The worn adjustment teeth were built up, and the combine was ready for work when it was needed

Fig. 4 After the corn was in the crib, this farmer took the time to have regular maintenance done on his corn elevator. By doing the work in the wintertime, the farmer was sure to have good equipment when he needed it

needed for work. Since the off-season was also the slow season for the local welding shop, the repairman was able to come to the farm to do the jobs, without interfering with other work in the shop. The farmer saved a trip to town, and the repairs to his equipment were made while he went about his normal work.

CONVERSION OF RAILROAD OBSERVATION CAR AIDED BY PLUG WELDING

Through an unusual application of plug welding, the Southern Pacific Co. recently expedited the application of steel sheets on both sides of a passenger car.

This fastening job came along in connection with converting an observation car which had been stripped to the bare framework into a new dome lounge passenger car. Three horizontal courses of 10-ft-long steel sheets, $^{3}/_{32}$-in. thick, were secured to the sides of the car. The lower and top rows, 4 ft and $1^{1}/_{2}$ ft wide, respectively, were of a low-alloy, high-strength, corrosion-resistant steel, while the 4-ft wide center row was smooth stainless steel. Later, corrugated stainless was secured to the lower and top rows. Channel members to which the sheets were plug welded were $^{1}/_{8}$-in.-thick, low-alloy steel, from 18 to 24 in. apart.

Before plug welding, the sheets were prepunched to fix the location of the studs on the framing members. Framing members meanwhile were cleaned of all paint with a sanding disk. A tack weld, to hold the sheets in position until plug welding was completed, was then made from the top edge of each sheet to a channel member. Plug welding with a stud welding gun was performed vertically, using mild steel studs for fastening the low-alloy sheets and stainless steel studs for the stainless sheets. Two stud-welding guns were used on the job, and a total of 6000 studs, $^{3}/_{8}$-in. in diameter, were plug welded.

An ingenious device was used to hold the sheets tight against the frame members—an essential condition for good plug welding. An 8-ft section of 1-in. pipe was fitted with a hook at the top and a sliding block of wood. The pipe was suspended vertically from horizontal frame members by the hook, and the wooden block was moved along the pipe to positions

Story contributed by Nelson Stud Welding Division of Gregory Industries, Inc., Lorain, Ohio.

Fig. 1 Plug welding of steel sheets to form the sides of a Southern Pacific dome lounge passenger car was facilitated by the ingenious device behind the stud welding gun in this picture. A sliding block of wood was moved along a vertical 8-ft section of pipe and used to press the sheets tight against the frame members

Fig. 3 Protruding ends of the $^{3}/_{8}$-in.-diam studs used for plug welding steel sheets, $^{3}/_{32}$ in. thick, to the framework of car, were pinched off with pliers. The small spaces between butted sheets were later filled by manual welding

close to the plug welds, where it was pushed against the sheets.

After the sheets were fastened, protruding ends of studs were pinched off with pliers. An emery disk was used on the stainless steel to remove excess material and to leave a polished appearance. The small spaces between butted sheets were filled with manual welding. There was no warpage in either the sheets or frame members.

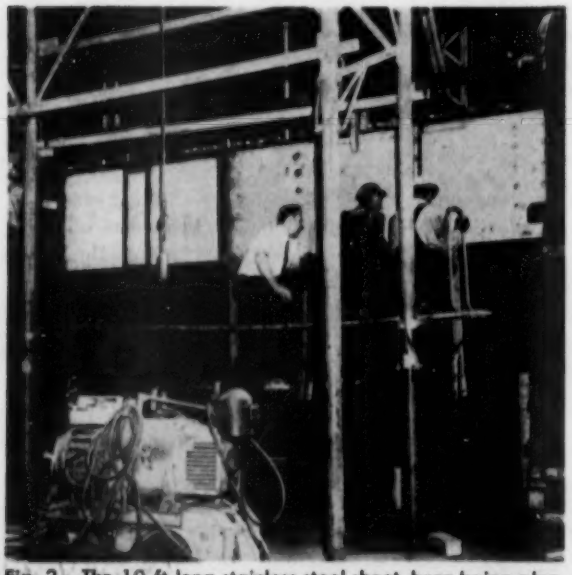

Fig. 2 The 10-ft-long stainless steel sheet, here being plug welded to the framework of car, was first prepunched to fix the location of the studs. Plug welding, which permitted covering both sides of the car in three days, left no distortion or warpage in either the sheets or the frame members

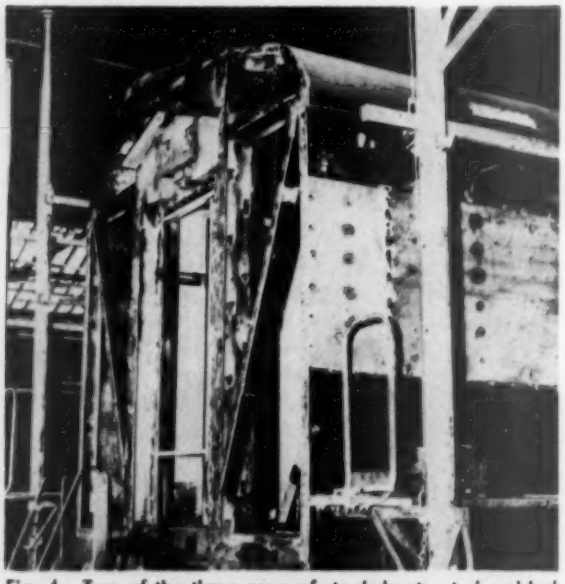

Fig. 4 Two of the three rows of steel sheets, stud welded to the framework of car, are shown here. The lighter sheet is smooth stainless, while the darker sheet below it is low-alloy steel, later covered with corrugated stainless. A narrower sheet of low-alloy steel was fastened along top

Society News

Papers on Current Practical Problems of Welding to Highlight National Spring Meeting in Kansas City

Papers on many current, practical problems of welding, as well as those on advanced research, will highlight the SOCIETY's program at the National Spring Meeting in Kansas City, June 7th to 10th. The tentative program, with a full listing of session topics, papers and authors appears in this issue of THE WELDING JOURNAL, page 254.

Greater emphasis is being placed on the Spring Meeting this year because of the recent revisions in the AWS By-Laws which changed the terms of the National Officers to begin and end in the Spring and which changed the date of the annual meeting to take place in the Spring.

As a result, it is becoming apparent that the time has arrived to establish the Annual Spring Convention, comprised of the AWS Welding and Allied Industry Exposition, National Spring Technical Meetings and the regular Annual Business Meeting as the Number One activity of the SOCIETY.

The Hotel Muehlebach in Kansas City, where the technical sessions will take place, is fully air-conditioned and thus will provide ideal facilities for the sessions. It is also close to the Municipal Auditorium where the annual Welding Show will be held.

The show, which will be the largest ever held in the welding field, will feature the greatest assemblage of welding equipment and accessories ever gathered under one roof. Present indications are that the original floor plans will need to be revised in order to accommodate all manufacturers who wish to exhibit. The present layout is already sold out.

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Although both conferences and exhibits are designed primarily for engineers and workers concerned directly with welding, the SOCIETY will this year invite students in vocational schools, engineering and agricultural colleges to inspect the displays. It is hoped that they will thus be exposed to the opportunities in the welding field and make some part of it their chosen vocation or profession.

The SOCIETY selected Kansas City as the site for the meeting and the show in line with its program to extend the frontiers of the welding science to all centers of American industry. The show in Kansas City will be the first heavy industry production show ever staged in that municipality.

Hotel reservations are now available and it has been decided to offer the choicest rooms on a first-come, first-served basis. Visitors who intend to come to Kansas City for the many welding events there should make hotel reservations promptly.

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Nine important sessions on new developments in welding are announced by the AMERICAN WELDING SOCIETY for the Ninth Western Metal Congress, March 28th–April 1st in the Ambassador Hotel, Los Angeles. In setting up the program a balance between heavy welding and aircraft—rocketry welding has been maintained.

Hugo W. Hiemke, president, California Alloy Products Co., Los Angeles, AWS program chairman for the congress, has announced that the sessions will be as follows:

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March 30th: A.M.—Welding of Pressure Vessels and Piping; P.M.—Heavy Plate Welding.

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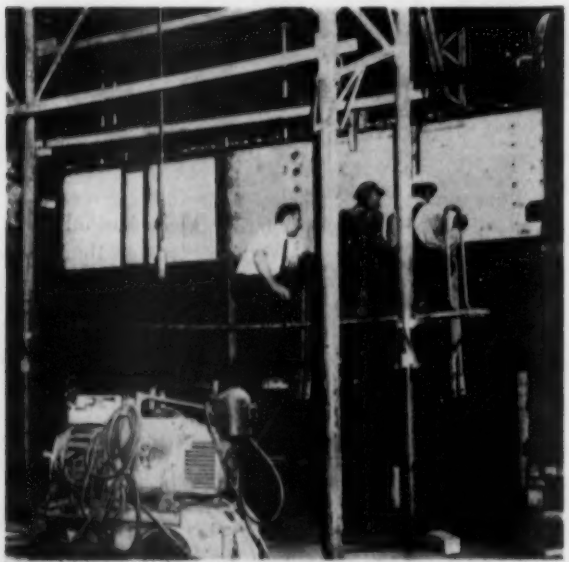

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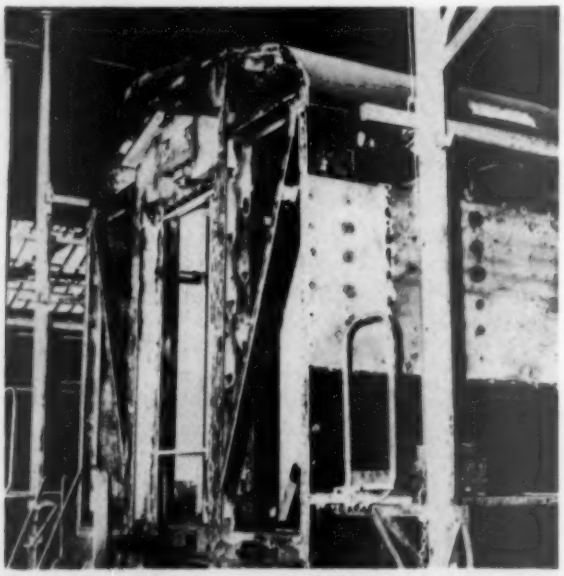

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PLANNING WELDING SESSIONS FOR WESTERN METAL CONGRESS

Four chairmen of AWS sessions compare program plans. *Left to right,* they are: Leo M. West, Francis H. Stevenson, Robert A. Hay and Arthur R. Wynn

Top men in Los Angeles Section discuss program plans for the Society's sessions. *Left to right:* Dr. Nathan A. Tiner, Byron R. Russell, Charles B. Smith and S. R. Lanier

Two leaders of AWS sessions in Ninth Western Metal Congress are shown consulting their parts in the technical program. *Left to right:* Russell Meredith and J. B. Arthur

Los Angeles, "Consumable Insert Ring for Stainless Steel Piping."

John H. Berryman, Air Reduction Sales Co., New York, "New Developments in Inert-Gas Welding."

Alan V. Levy, supervisor, material and process section, and Robert W. Wickham, welding engineer, Marquardt Aircraft Co., Van Nuys, Calif., "Inert Gas Welding of Titanium."

R. W. Tuthill, General Electric Co., York, Pa., "New Techniques in Inert-Gas-Shielded Metal-Arc Welding."

The Brazing and Hard Facing session will have for chairmen: George Griffin, Victor Equipment Co., Los Angeles, and Alex Maradudin, materials engineer and metallurgist, Standard Oil Co., of California, El Segundo.

Speakers and topics for this session are:

C. H. Chatfield, Handy and Harman Co., Bridgeport, Conn., "Silver Brazing of Refractory Metals."

George C. Driscoll, Western Alloy Engineering Co., Montebello, Calif., "High-Temperature Alloy Brazing."

Otis Cullum, Bakersfield, Calif., "Hard Facing of Oil Well Drill Pipe."

The Aircraft-Rocketry session for March 29th will have for chairmen: Richard C. Hayes, process materials engineer, Douglas Aircraft Co., Long Beach, Calif., and Harlan L. Meredith, engineer welding research, North American Aviation, Inc., Downey, Calif., both of whom at Mr. Hiemke's request made up the program. It will consist of outstanding welding authorities engaged in aircraft and rocket work, as follows:

Leon Shafin, superintendent, brazing, Gilfillan Bros., Los Angeles, "Silver Alloy Brazing of Electrical Connections."

D. O. Samuelson, welding engineer, and F. G. Harkins, chief welding engineer, Solar Aircraft Co., San Diego, "Resistance Welding of Airframe Structures."

Charles W. Handova, metallurgist, and Harlan L. Meredith, of North American, Downey, "Titanium Alloy Fusion Weldability and Correlated Weld Metallurgy."

Byron Russell, owner-manager, Airline Welding and Engineering, Hawthorne, Calif., "New Full Automatic Weld Tooling for Aircraft and Missile Joinery."

Nathan A. Tiner, research engineer, North American Aviation, Downey, "Metallurgical Aspects of Silver Brazing of Titanium."

Sander Appointed to Technical Council

C. P. Sander, general superintendent of the Consolidated Western Steel Division, U. S. Steel Corp., Vernon, Calif., has been appointed a member of the Technical Council of the Board of Directors of the AMERICAN WELDING SOCIETY. Mr. Sander held the office of director-at-large of the SOCIETY from October 1953 to November 1954. He was nominated and elected to the office of second vice-president of the AMERICAN WELDING SOCIETY in 1954 and assumed that office on Nov. 4, 1954, for a term extending to May 31, 1956. Hei s also a member of the SOCIETY's executive committee.

The Technical Council is composed of leading authorities in the welding industry and it is charged with the supervision, planning and expediting of all technical affairs of the Society.

Cleveland Section Has Spot on TV Program

"Welding Your Future" was the title of an unusual program presented over TV station WNBK on Sunday, Jan. 16, 1955, from 12:30 to 1:00 P.M., by the Cleveland Section of the AMERICAN WELDING SOCIETY.

The program was sponsored by the Cleveland Technical Society's Council as "Adventures in Engineering & Science."

Participating as a representative of the graduating class of Greater Cleveland was Larry Noble of the senior class of Euclid High School, who asked questions regarding different phases of the welding field before a panel consisting of several engineers, Charles Herbruck of the Lincoln Electric Co., Ross Yarrow, superintendent of Republic Structural Iron Works Division, Wasil Romance of Euclid Division of GMC, and William Mumford of Diebold Engineering Sales Co., all of whom are members of the Cleveland Section.

Through their explanations and a film entitled "The Magic Wand of Industry," viewers saw actual welding operations and the contributions welding has made to the industry. Educational requirements for welding engineering and the opportunities awaiting young men in the welding field were pointed out.

Most of the material for this program was supplied by Cleveland industries. It is understood that this will be a regular affair, with the Cleveland Section participating in approximately two programs a year.

VISIT THE
WELDING SHOW
MUNICIPAL AUD.
Kansas City, June 8–10

NOW! ONE WELDER

IDEALARC... GIVES YOU

BOTH AC and DC

WELDING CURRENT

NOW
DC Industrial Welder for operation on single phase power lines

NOW
Rectifier Welder with both current and voltage controls

NOW
Rectifier Welder with arc-booster starting

● One machine, Lincoln Idealarc, now gives you *both* AC and DC welding current. You have *both* current control and voltage control. For every job, you can now select the ideal arc . . . soft arc or forceful arc . . . AC or DC.

Shops which do not have 3 phase power can now use DC. Idealarc is the only DC welder to operate from single phase power available at reasonable prices in industrial sizes.

With dual arc control on DC, you have the same benefits of soft arc and forceful arc for welding with DC as with AC. Arc-booster starting on both AC and DC . . . on both soft arc and forceful arc . . . assures non-sticking, easy operation, speeds intermittent welding, gives full penetration at the start of each weld.

Idealarc cuts duplication of welding equipment . . . saves on welder cost. So why tie down your operations with one choice of welding current? Idealarc provides complete freedom to select the correct arc type and welding current for maximum speed, maximum ease and quality of welding . . . for the same investment.

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Send for Bulletin 1343
Shows how you can cut welding costs . . . now. Write:

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THE WORLD'S LARGEST MANUFACTURER OF ARC WELDING EQUIPMENT

TECHNICAL PAPERS SESSIONS

1955 NATIONAL SPRING MEETING

JUNE 7-10 • MUEHLEBACH HOTEL, KANSAS CITY, MO.

JUNE 7, TUESDAY AFTERNOON

Three Simultaneous Sessions

1. INERT ARC WELDING

A. Some Fundamentals of Inert Gas Arc Welding
by H. C. Ludwig, Westinghouse Electric Corp.

B. Electrode Activation for Inert-Gas-Shielded Arc Welding
by Alexander Lesnewich, Air Reduction Co., Inc., Research Labs.

C. Inert-Gas Welding in the Aircraft Industry
by John M. Thompson, Jr., CONVAIR

2. RESISTANCE WELDING

A. The Flash Welding of Commercial Molybdenum
by E. F. Nippes and W. H. Chang, Rensselaer Polytechnic Institute

B. The Static and Fatigue Behaviors of Spot-Welded Joints in Titanium
by W. H. Kearns, W. S. Hyler and D. C. Martin, Battelle Memorial Institute

C. Oscillographic Instruments in Spot Welding Quality Control and Maintenance
by G. C. Woodmancy, Boeing Airplane Co.

3. WELDABILITY AND RESEARCH

A. Effect of Welding on Transition Temperature of Nickel Steel Plate
by T. N. Armstrong, International Nickel Co., and W. L. Warner, Watertown Arsenal

B. Initial Characteristics of Chromium-Nickel Steel Weld Metals
by J. Heuschkel, Westinghouse Electric Corp.

C. Properties of Stainless Steel Welds at Ambient and −40° F Temperatures
by V. N. Krivobok, International Nickel Co., and A. Choquet and G. Welter, École Polytechnique

JUNE 8, WEDNESDAY MORNING

Three Simultaneous Sessions

4. INERT ARC WELDING

A. Carbon-Dioxide-Shielded Consumable-Electrode Arc Welding
by Gilbert R. Rothschild and Harry C. Cook, Air Reduction Co.

B. A Straight Polarity, Inert-Gas Process for Welding Mild Steel
by J. M. Cameron and A. J. Baeslack, Westinghouse Electric Corp.

C. Consumable-Electrode Inert-Arc Spot Welding
by R. L. Hackman, Linde Air Products Co.

5. WELDABILITY AND RESEARCH

A. Weld-Metal Flaw-Evaluation Studies
by R. S. Ryan and P. J. Rieppel, Battelle Memorial Institute

B. Investigation of the Importance of Quench Aging in ASTM A-300
by E. F. Nippes, Rensselaer Polytechnic Institute, and W. Fleischmann, Knolls Atomic Power Laboratory

C. Investigation of Weld-Metal Cracking in High-Strength Steels
by R. P. Sopher and P. J. Rieppel, Battelle Memorial Institute, and A. J. Jacobs, Ohio State University

6. STRUCTURES

A. Static Tests of Arc-Welded Aluminum Alloy Beams
by M. Holt and R. B. Matthiesen, Aluminum Company of America

B. Welding of Plate Girder Bridge by Submerged Arc Welding
by J. A. Hoffman, Bethlehem Steel Co., Shipbuilding Div.

C. Maintenance Welding and Cutting Operations on Radioactive Process Equipment
by E. B. LaVelle and J. M. Fox, Jr., General Electric Co.

JUNE 8, WEDNESDAY AFTERNOON

Three Simultaneous Sessions

7. SYMPOSIUM ON QUALIFICATION

Sponsored by Technical Activities Committee.

A. Basic Philosophy

B. Application to Arc and Gas Welding

C. Application to Resistance Welding

D. Application to Brazing Processes

E. Application to New Welding Processes

Speakers to be announced at a later date.

8. AIRCRAFT AND ROCKETRY

A. Fusion Welding Unalloyed Titanium Sheet Without Filler Metal
by A. V. Levy and R. Wickham, Marquardt Aircraft Co.

B. Titanium Alpha and Alpha-Beta High Strength Alloy Fusion Weldability
by H. L. Meredith and C. W. Handova, North American Aviation, Inc.

C. Inert-Gas Consumable Electrode Arc Welding of 4130 Steel
by C. R. Sibley and D. R. Kelker, Air Reduction Sales Co.

9. PROCESSES

A. Surfacing with Composite Tube Rod
by R. P. Culbertson, Haynes Stellite Co.

B. Evaluation of Fuels and Oxidants for Welding and Associated Processes
by Walter B. Moen and John Campbell, Air Reduction Co., Inc.

C. High-Speed Welding of Light-Gage Material
by R. A. Kubli and T. McElrath, Linde Air Products Co.

JUNE 9, THURSDAY MORNING

Two Simultaneous Sessions

10. PRESSURE VESSELS

A. Properties and Characteristics of a Quenched and Tempered Steel for Pressure Vessel Construction
by W. D'Orville Doty, United States Steel Corp.

B. Pressure Vessel Design Using High-Strength Quenched and Tempered Steel
by L. P. Zick, Chicago Bridge & Iron Co.

C. Suitability of Quenched and Tempered Steels for Pressure Vessel Construction
by Leon C. Bibber, United States Steel Corp.

11. WELDING ELECTRODES

A. Hardenability Evaluation of Welding Electrodes
by Leo M. West, Douglas Aircraft Co.

B. Evaluating the Iron Powder Coated Electrodes for Production Use
by D. B. Howard, ACF Industries, Incorporated.

C. The Effect of I^2R Heating on Electrode Melting Rate
by J. L. Wilson, G. E. Claussen and C. E. Jackson, Metals Research Labs., Linde Air Products Co.

JUNE 10, FRIDAY MORNING

Two Simultaneous Sessions

12. PIPING AND PIPE LINES

A. Inert-Gas Tungsten Arc Welding of Piping
by H. Thielsch and C. S. Pulliam, Grinnell Co., Inc.

B. Stringer Bead Welding of Carbon Steel and Low-Alloy Pipe by Inert-Gas-Arc Welding
by J. Neal, The Fluor Corp.

C. Temper Brittleness of 1 Cr-$^1/_2$ Mo Weld Metal
by W. J. Lester and G. R. Prescott, C. F. Braun & Co.

13. BRAZING

A. Lithium Additions to Brazing Alloys
by N. Bredzs and D. A. Canonico, Armour Research Foundation

B. A Metallurgical Study of Silver-Cadmium-Copper-Zinc Brazing Alloys
by K. M. Weigert, Goldsmith Bros. Smelting & Refining Co.

C. Semi-Automatic Multiple Flame Brazing Larger Brass Electronic Components
by J. W. Weyers and L. E. Mills, General Electric Co.

The above schedule of technical papers is not necessarily final. A more complete and detailed program, including an account of other activities, will be published in the May issue of The Welding Journal.

Section News and Events

as reported to Catherine O'Leary

STAINLESS STEEL

Albuquerque, N. Mex.—The *Albuquerque* Section held its January meeting on the 21st at the Industrial Arts University. Virgil C. Reed (M), of the Alloy Rods Co., spoke on the subject "Stainless Steel" and showed a film on the manufacture and testing of alloy rods. The film was supplied by the Alloy Rods Co.

FABRICATING PROCEDURES

Fort Wayne, Ind.—The January 20th meeting of the *Anthony Wayne* Section was held at Howard Johnson's Restaurant in Fort Wayne, with 32 members and guests present.

Chairman Jacob Geiser introduced the speaker, Edward E. Beyer (M), supervisor for the R. C. Mahon Co., Detroit, Mich.

Mr. Beyer gave an excellent illustrated talk on "Fabricating Procedure in a Job Shop." He also showed a Westinghouse sound film entitled "A-C vs. D-C Welding."

ECONOMICS AND EDUCATION

Phoenix, Ariz.—The two speakers at the Dec. 19, 1954, meeting of the *Arizona* Section were well qualified in their presentation of their particular subjects.

Hal Savage (M), of the Arizona Welding Equipment Co., discussed "Profit Sharing" as related to the welding industry and the possibilities of expansion under an incentive plan.

Ralph Smith (M), of Smith & Klemmer Co., gave a fine talk on the necessity of education within the Section so as to widen the scope of welding knowledge and present it to a greater number of people.

These talks were followed by a presentation of a color film "Cerro Bolivar." This is an educational film, although not specifically related to welding. It described the finding and exploiting of vast iron ore deposits in Venezuela by the United States Steel Corp.

WELDING DEVELOPMENTS

Boston Mass.—The *Boston* Section held its regular monthly meeting on Monday, January 10th, at the Westinghouse Auditorium.

In place of the usual coffee speaker, an excellent film on fishing, "Wings for the Angler," was shown.

The technical speaker was Allen G. Hogaboom (M), Bethlehem Steel Co. Shipbuilding Division, Quincy, Mass. Mr. Hogaboom talked about "New Developments in the Welding Field." He covered the newest developments in welding abroad as well as domestic advances in electrode research, automatic welding power sources, elimination of high frequency interference in inert-gas-shielded arc welding and other modern processes.

ALLOY STEEL PLATE

Boston, Mass.—On Monday evening, Dec. 13, 1954, the *Boston* Section held its regular meeting at the Westinghouse Auditorium.

After the usual social hour, about 92 members and guests enjoyed another of Dora Ferguson's outstanding dinners.

The Coffee Speaker of the evening was H. Russell Beatty, president of

SECTION MEETING CALENDAR

APRIL 1st

NORTH CENTRAL OHIO Section. Dinner meeting. 6:30 P.M. Mansfield, Ohio.

APRIL 11th

DAYTON Section. Joint Meeting with ASTE. Sutmiller's Restaurant. "Automation" by C. F. Hautau, Chief Eng., Hautau Eng. Co.

APRIL 12th

NEW YORK Section. 6:00 P.M. Schwartz's Restaurant, 54 Broad Street, NYC.

WESTERN MASSACHUSETTS Section. Springfield, Mass. Movies: "Canadian Pattern, Canada's Tackle Busters."

APRIL 14th

NORTHERN NEW YORK Section. Pittsfield, Mass. "Position All Three. The Man, The Weldment, The Automatic Head," by A. K. Pandjiris, Pandjiris Weldment.

Technical Activities Display—"Review of Categories and Ordering Procedures."

SAGINAW VALLEY Section. Dinner Meeting. High Life Inn, Saginaw, Mich. "Tool and Die Welding," by R. K. Lee, Alloy Rods Co.

APRIL 18th

ROCHESTER Section, Rochester, N. Y. Program of papers submitted by local members.

APRIL 19th

NEW JERSEY Section. Dinner 6:30 P.M. Essex House, Newark. "Practical Fabrication."

RICHMOND Section, Richmond, Va. "Building Structure and the Importance of Saxe System Welded Connection Units for Welded Assemblies," by Van Rensselaer P. Saxe, Engineer. Baltimore, Md.

APRIL 20th

NORTHWESTERN PENNSYLVANIA Section. Erie, Pa. "Product Design for Welding," by J. M. Mikulak.

APRIL 22nd

MILWAUKEE Section. "High-Speed Submerged-Arc Welding of Mild Steel with Multiple Arcs," by W. H. Schuster.

APRIL 29th

PITTSBURGH Section, Pittsburgh, Pa. Georgian Room, Webster Hall Hotel. Eighteenth Annual Tri-State Ladies' Night and Dinner.

Editor's Note: Notices to be published in any one given issue must reach Journal office not later than the first of the preceding month. Give full information concerning time, place, topic and speaker for meeting.

Wentworth Institute, Boston. The title of his talk was: "Opportunities for the Engineering Technician in Metallurgy in New England." As part of his presentation, Mr. Beatty inquired as to the need for a training program for welding technicians, to fill a possible gap between the College trained welding engineer and the shop level. Mr. Beatty indicated that he would appreciate hearing from the members at large on this point.

The Technical speaker was R. C. Altman of the Metallurgical Division, U. S. Steel Corp., Pittsburgh, Pa. His subject was a very interesting and timely one dealing with the new T-1 Alloy Steel Plate, which develops a minimum yield point of 90,000 psi. The first part of the program illustrated with colored slides the metallurgical and welding characteristics of this alloy, which can be welded without either preheating or postheat treatment. The balance of the talk was accompanied by a movie, showing tests of tanks 4 ft in diameter and 20 ft long at temperatures well below zero. The tanks were tested to destruction both by the application of hydraulic pressure to bursting, and the dropping of a 13-ton weight on the tanks, while under pressure and supported at the ends, from a height of 100 ft.

ELECTION OF OFFICERS

Chattanooga, Tenn.—The *Chattanooga* Section announces the election of the following officers for the fiscal year:

Chairman—Roy Mitcham, Hotel Patten, Chattanooga, Tenn.
1st Vice-Chairman—William H. Barker, Linde Air Products Co., Chattanooga, Tenn.
Secretary—Peter S. Naar, American Bureau of Shipping, 614 James Building, Chattanooga, Tenn.
Treasure—Raymond E. Barrett, Brooks Welding Supply Co., 1200 Dayton Blvd., Chattanooga, Tenn.
Chairman, Program Committee—George E. Tepley, C. D. Center Co., Chattanooga, Tenn.

RESISTANCE WELDING SYMPOSIUM

Chicago, Ill.—On January 21st well over a hundred members and guests of the *Chicago* Section attended a Resistance Welding Symposium. Jerome Welch (M'52), process and manufacturing engineer of Cutler Hammer, Inc., showed some very interesting new modifications of spot-welding equipment. John J. Chyle, director of welding research, A. O. Smith Corp., and first vice-president of AWS, concluded the technical session. Mr. Chyle spoke on flash welding applications and the metallurgical considerations involved.

Section Chairman Amel Meyer introduced A. F. Chouinard, of National Cylinder Gas Co., who is replacing Mr. Welch as district director.

Bill Craske, of Arcos Corp., did his usual outstanding job as chairman of the Host Committee. Bill and his group are really making guests feel most welcome.

Felix Tancula of the *Welding Engineer* has been appointed official photographer of the Section.

Jay Bland, of Standard Oil, reviewed the membership situation. He reviewed the activities of his group and outlined plans for increasing membership. Ed Zahumensky of Taylor Forge and Pipe Works is a very active member of the Membership Committee.

O. T. Barnett reported that advance registration for the Midwest Welding Conference, sponsored jointly by the Chicago Section and Armour Research Foundation, assured a tremendous turnout.

Have You Made
Your Reservations
For Kansas City?

AT THE RESISTANCE WELDING SYMPOSIUM IN CHICAGO

This is part of the large crowd in attendance at the dinner which preceded the Resistance Welding Symposium held by the Chicago Section on January 21st

The principal speakers of the evening were Jerome Welch (*left*) and John J. Chyle (*right*)

WELDING AND CUTTING

Cincinnati, Ohio.—The January 25th dinner meeting of the *Cincinnati* Section was held in the Engineer and Scientist Building. The excellent dinner was highlighted by the witty comments of Cincinnati's own Dick Bray. The questions on sports fired at Dick were strictly local and were received with vivid enthusiasm.

The big question today is "What's With This CO_2 Welding?" The Cincinnati Section received the answer to this question first hand from the man who contributed much to its development—Edward Vogel, research engineer for the Liquid Carbonic Corp. Not only did Mr. Vogel explain the operation but he also demonstrated the new process. The use of CO_2 in combination with other inert gases still is in its infancy. The Section members were fortunate in obtaining an insight on this current development.

Gases do play a very important part in the field of welding. Flame cutting itself is not new but the struggle for improvement still goes on. R. F. Helmkamp (M'52), Air Reduction Co., pointed out the technical aspects

of the design of the oxygen orifices in cutting torches. The cost of templets used in cutting out intricate patterns with accurate results, is quite high. Considerable savings can be realized by laying out the shape to be cut on masonite and using a photoelectric cell to do the tracing. Examples of this cost reduction were taken from actual cases. The cutting of bevels on irregular shapes has proved to be a complicated task. A development of a mechanism to do the job with ease and accuracy was presented by Mr. Helmkamp.

"PAY DIRT"

Cleveland, Ohio.—Feature speaker at the January Dinner Meeting of the *Cleveland* Section held at the Mange Hotel on January 12th was Upshur Evans, executive director of the Cleveland Development Foundation. Mr. Evans' talk entitled "Pay Dirt" concerned the Foundation, what it is, what its plans are for the development of Cleveland and Northeastern Ohio, and what it will mean to the welding industry and to business and industry in general in this area in the future.

ANNUAL LADIES' NIGHT

Denver Colo.—The December meeting of the *Colorado* Section held on Dec. 14, 1954, was social in order to entertain the wives and the members. An extra nice dinner, with table flowers and favor corsages for the ladies, was one of the attractions of the evening. An after dinner feature was the showing of a very interesting Disney picture of the Beaver Country in color.

LOW-HYDROGEN ELECTRODES

Denver, Colo.—The *Colorado* Section met on January 11th at the Oxford Hotel. A movie on "Steel" was shown before the technical meeting. This movie was of general interest showing the process of reducing steel billets to woven steel fabric ranging from reinforcing for concrete to poultry netting on the farm.

The technical speaker was Richard L. Lee [illegible], vice-president, Alloy Rods Co. Mr. Lee gave an unusually interesting and illuminating talk on low-hydrogen electrodes. Mr. Lee is an excellent speaker and knows a lot of useful answers to questions about welding rods.

MULTISPHERES

Columbus, Ohio.—The design, development and construction of multispheres was the subject of a talk by Christian Arne of the Chicago Bridge and Iron Co., given at the January 14th meeting of the *Columbus* Section. Mr. Arne is well qualified to discuss this subject since he had much to do with the original design and development of the first multispheres. He is author of "New Method in Spherical Geometry" and contributed to the third edition of the AWS WELDING HANDBOOK. He has also been granted a number of patents. Mr. Arne has been associated with Chicago Bridge since 1937, with the Erection Department until 1941 when he transferred to the Shipbuilding Division. Since 1945 he has been a research engineer.

Christian Arne

Continuing the Section's series of films on the subject of atomic energy, two were shown after the meeting. They were Civil Defense films on "Operation Greenhouse" and "The House in the Middle."

QUIZ PROGRAM

Dayton, Ohio.—The January meeting of the *Dayton* Section was in the form of a welding quiz program with a panel. The regular meeting began at 8 P.M. at the Dayton Engineers Club. A movie entitled "The Soviet Union and Its People" was shown at the beginning of the meeting, after which the 50 questions, multiple choice quiz was given to everyone present. This quiz covered resistance, inert gas and acetylene welding as well as coated electrodes and brazing. After the quiz had been completed by everyone, the papers were exchanged and the correct answers given and discussed by the panel of experts, consisting of Charles O. Adams, Robert C. Appenzellar, S. E. Garber and Herbert G. Weiler. Glyn Williams won first prize, Bob Bowman won second and there was a tie for third between Harry Jones and Carl I. Bertsche. The prizes were a Troy Sunshade reclining chair, a lightweight traveling bag and a four-person coffee service.

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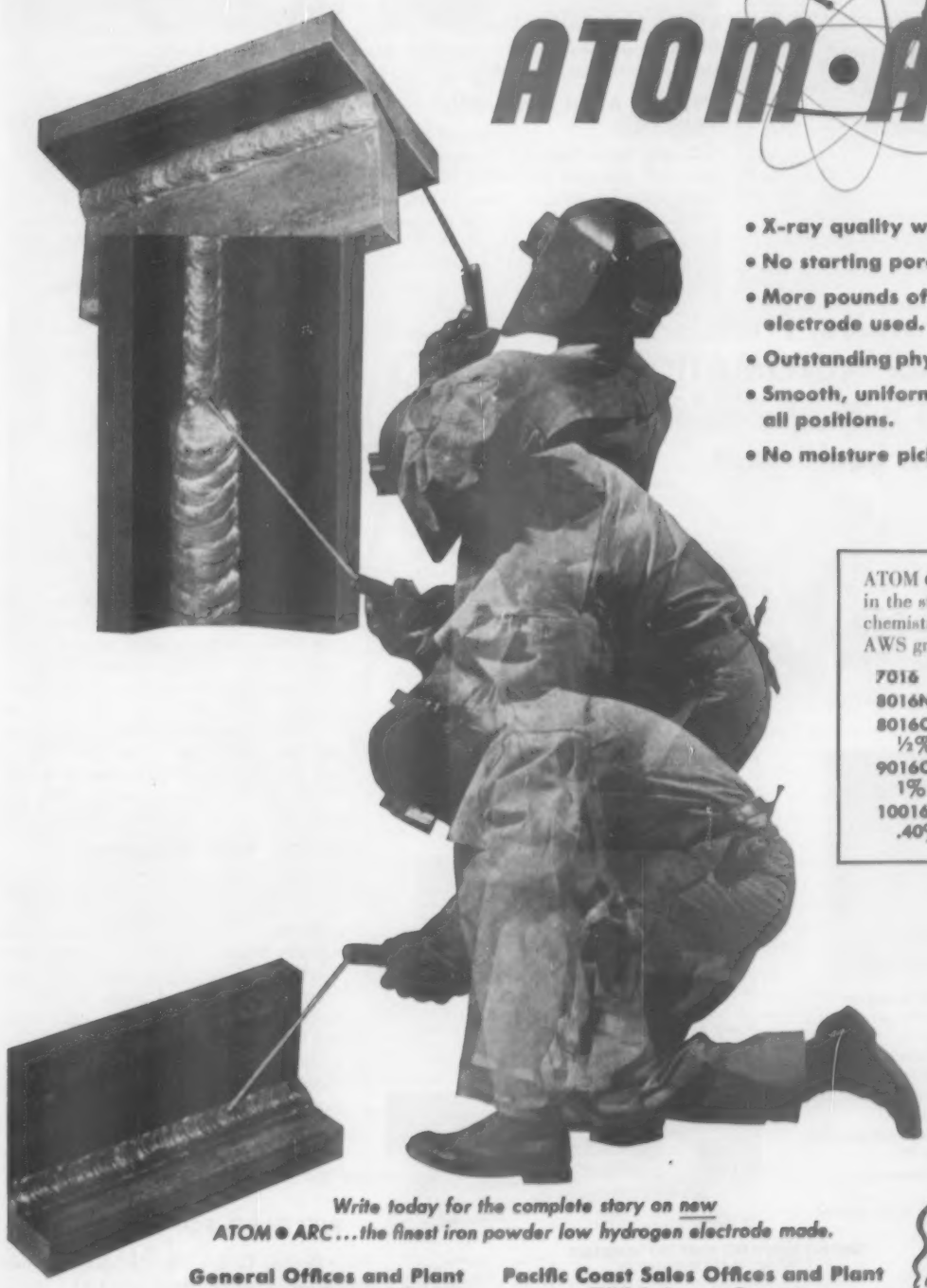

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CHRISTMAS PARTY

Appleton, Wis.—A Christmas Party for the *Fox Valley* Section members, their friends and wives was held at the Elks Club on Dec. 10, 1954. Cocktails were served after which approximately 110 people sat down to a splendid dinner. After dinner there was a visit from "Santa Claus" who brought appropriate gifts for the ladies. The evening was concluded with dancing from 9:00 until 1:00.

Two typical scenes of the merrymakers at the Christmas Party held by the Fox Valley Section on December 10, 1954

RESISTANCE WELDING

Appleton, Wis.—The regular monthly meeting of the *Fox Valley* Section for January was held on the 21st at the Elks Club. Speaker was A. L. Williams, of the Federal Machine & Welder Co. Mr. Williams presented an excellent address on "Resistance Welding—Uses and Applications."

STRESS ANALYSIS AND TESTING

Des Moines, Iowa.—The *Iowa* Section held a panel discussion of stress analysis and testing on Nov. 18, 1954, at the Hotel Kirkwood. The moderator of the discussion was Prof. L. S. Linderoth, Mechanical Engineering Department, Iowa State College. Members of the Panel were: M. L. Patzig, Patzig Testing Laboratories, Des Moines; Karl Henderickson, Link Belt Speeder Co., Cedar Rapids; L. H. Traviss, John Deere Des Moines Works; and W. O. Boian, Pittsburgh Des Moines Steel Co. Questions were directed to the panel from the floor.

Preceding the panel discussion, Iowa State Conservation Officer Lloyd Huff gave a short but interesting résumé on Iowa Game Laws and also described many of Iowa's state parks and the facilities available at each.

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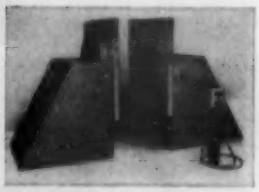

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CHRISTMAS PARTY

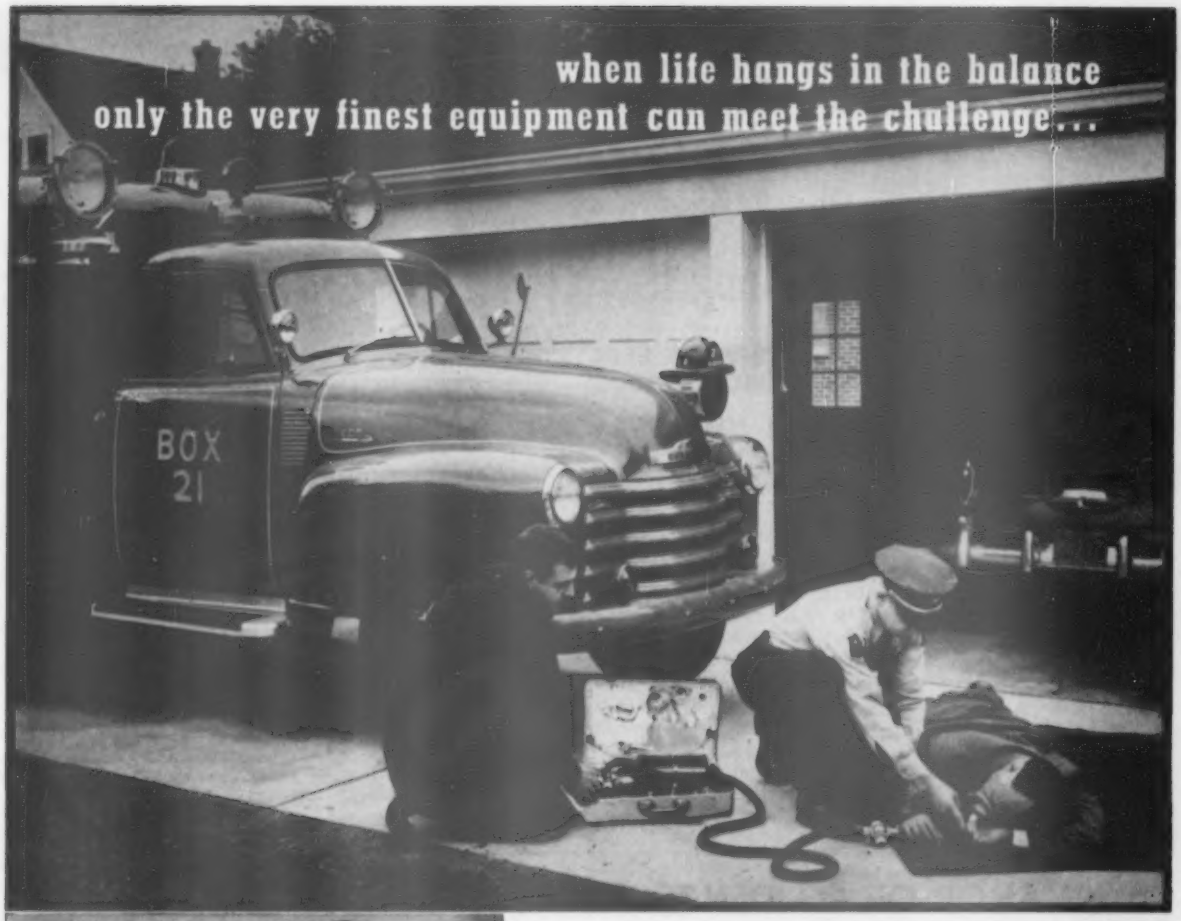
Des Moines, Iowa.—The annual Christmas party of the *Iowa* Section was held on the evening of Dec. 16, 1954, at the Hotel Kirkwood. A thick steak dinner was the feature attraction. Several door prizes, card playing and movies rounded out the evening's entertainment. Many new friends and lots of old friends thoroughly enjoyed themselves, making this year's party the best yet in the three years of the existence of the Section.

INERT ARC WELDING

Moline, Ill.—A discussion of the application of inert arc welding was presented by P. G. Parks, chief welding engineer of the Solar Aircraft Co. at the January 13th meeting of the *Iowa-Illinois* Section. Mr. Parks illustrated with slides and other data the equipment used in production of weldments of jet aircraft. The speaker also devoted considerable time to a question and answer forum on the subject and was presented a money clip as a token of appreciation for his service.

WELDING FORUM

Hollis, L. I.—The January Dinner and Technical Meeting of the *Long Island* Section, held on the 13th, featured a welding forum covering fusion,



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brazing, spot and flash-butt welding, inert-gas-shielded arc welding, both consumable and tungsten electrode methods on aluminum, mild, alloy, stainless and corrosion-resistant steels. Forty hardy souls braved the icy roads to attend and take part in the forum and general discussion held at Anselmi's Restaurant, Bethpage, L. I.

Moderator for the evening was Anton Anguilo, Nassau Suffolk Welding, Inc., Bellmore, L. I. Panel members were: Henry Hoffman, welding engineer, Todd Shipyards, Brooklyn; Bill Kollner, welding superintendent, George Malvese Co., Inc., Garden City Park, L. I.; Earl Morris, assistant chief tool engineer, Metals and Processes, Grumman Aircraft Engineering Corp., Bethpage, L. I., and Al Strickland, Linde Air Products Co., New York.

The popularity of this type of meeting was attested to by the general participation of those attending and the number of requests for more of the same. The usual social period preceded the meeting.

ALUMINUM DROP TANKS

Los Angeles, Calif.—The Aircraft and Rocketry Welding Panel of the *Los Angeles* Section held its January meeting on the 6th at Swally's Cafe. In place of a coffee speaker, there was a picture of a fishing and hunting trip made by Johnny Tolan of Research Welding and Byron Russell of Airline Welding into the great Teton Mountains and the Jackson Hole Country of Wyoming.

The speaker of the evening was Byron Russell [illegible], of Airline Welding. He gave a very excellent talk illustrated by 3-D color slides of the latest developments in tooling and jig fixtures for the automatic welding of aluminum drop tanks.

BRONZE ELECTRODES AND FILLER RODS

Madison, Wis.—The regular meeting of the *Madison* Section was held on Dec. 29, 1954, at the Playdium. Dinner and a short business meeting preceded the technical program. Technical speaker was F. Emery Garriott [illegible], Weldrod and Wire Department Manager, Ampco Metal, Inc., Milwaukee, Wis. Mr. Garriott's subject was "Jobs You Can Do with Bronze Electrodes and Filler Rods." The every-day usage of bronze electrodes and filler rods in the joining of iron base metals was discussed in addition to their use with dissimilar methods overlaying for wear and corrosion resistance, and other interesting applications.

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NEWLY ELECTED OFFICERS

Madison, Wis.—The following officers have been elected by the *Madison* Section for the fiscal year:

Chairman—Walter Griskavich, 206 S. Midvale Blvd., Madison 5, Wis.

1st Vice-Chairman—Howard Adkins University of Wisconsin, Madison 5, Wis.

Secretary - Treasurer—Fred E. Theiler, Madison Vocational and Adult School, 211 North Carroll St., Madison 3, Wis.

WELDING OF STEELS

Baltimore, Md.—The 105 hardy souls who braved the cold on January 21st to attend the monthly meeting of the *Maryland* Section at the Engineers Club were well rewarded for their efforts by hearing a most enlightening talk by George Linnert [illegible], research welding metallurgist for the Armco Steel Corp. The subject of Mr. Linnert's talk was "Is the Welding of Steels Art or Science?"

Mr. Linnert very capably demonstrated that welding is not an art, but that it is a science, which calls for some knowledge of various branches of science, such as metallurgy, chemistry, physics, strength of materials, etc. Numerous photos were projected which showed examples of failures in welded structures; failures in the form of rupture of the joints, or by corrosion, all of which were due to improper procedures as a result of a lack of certain scientific knowledge.

It was explained how the "artist" frequently gets into serious difficulty because he likes the way certain electrodes of his choice "behave," and the speaker told how many serious corrosion problems were solved by careful scientific methods.

There can be little doubt that all who heard this talk are convinced that welding is a science.

Many of the past chairmen were present at this meeting, including the following: Van Rensselaer P. Saxe, 1932–33; Bela Ronay, 1934–35; Jack Cumberland, 1944–45; Wes Shaprow, 1945–46; Al Earlbeck, 1947–48; George Linnert, 1948–49; E. B. Lutes, 1949–50; Fritz Albrecht, 1950–51; Vic Lees, 1951–52, and Harrison Sayers, 1952–53. There were also a number of out of town visitors at this meeting.

LADIES' NIGHT

South Bend, Ind.—The *Michiana* Section held a Ladies Night event on January 20th. After a few welcoming remarks by the chairman, a delicious country style dinner was en-

joyed by all. The table was decorated with red roses and each lady was given a rose at the conclusion of the meal.

Presentation of a Past-Chairman Pin was made to Erwin Laylin, who was chairman during the 1953–54 season.

The entertainment part of the evening was ably handled by Mr. and Mrs. A. H. Rerick of South Bend, whose subject was "Echoes of India." A large number of colored slides taken during their trip to India were explained in detail by Mr. Rerick. Mrs. Rerick exhibited a beautiful gown made of Indian material and assisted in outfitting a lady from the group, who served as a model, in a complete Indian custom. Articles of jewelry and objects of art were also shown.

A very pleasant social evening was enjoyed by all.

ISOTOPES AND INSPECTION

Newark, N. J.—A joint meeting of the *New Jersey* Section with the ASME was held on Tuesday, Jan. 18, 1955, at the Essex House in Newark. Dinner preceded the technical meeting which started at 8:00 P.M. with an attendance of 200 members and guests.

The speaker of the evening was Robert Epple, supervisor of the Technical Service Division, Tracerlab, Inc. His well-illustrated talk covered the use of various isotopes and their applications to the radiographical field; also, the use of isotopes as an inspection tool to the welding industry, and the availability of these isotopes.

TOOLING FOR AUTOMATIC WELDING

Newark, N. J.—The December meeting of the *New Jersey* Section was held on Tuesday, Dec. 14, 1954. Dinner was followed with the showing of a short film entitled, "Big Scot." This film showed the raising and training of Cylesdale Brewery horses.

The meeting started at 8:00 P.M. and was attended by 120 members and guests. The speaker, Anthony K. Pandjiris (M'52), president of Pandjiris Weldment Co. discussed problems and solutions connected with manipulation tools for automatic welding; also, advantages, disadvantages and examples. He also gave some information on costs.

NIETZEL AWARD

New York, N. Y.—The meeting of the *New York* Section held on January 11th at Schwartz's Restaurant was the most successful one of the current season thus far. In addition to a good attendance, four national officers of the SOCIETY were present, namely, President J. H. Humberstone, Secretary J. G. Magrath, Assistant Secretary F. J. Mooney, and Membership Chairman E. Dato. The purpose of their presence was to present a membership award to the New York Section.

At this meeting, J. Mikulak (M'52), of the Worthington Corp., presented a paper on powder metallurgy and the use of iron powder for electrode coatings. The Worthington Corp. has made extensive tests in this field and Mr. Mikulak conveyed his company's findings to the Section. The talk was well received and resulted in numerous questions from the audience.

W. Lukas of the Taylor Forge and Pipe Works, chairman of the New York Section, introduced A. Seiden, architect, chairman of the Section Membership Committee, who gave a brief summary of the Committee's activities during 1954. Mr. Seiden attributed the success achieved during the year to the good assistance given by the committee members, particularly by W. Kriewall of the American Agile Corp. Mr. Seiden did not comment on his own efforts, but it is well known that his personal contribution to the bringing in of new members to the Section has been considerable.

Chairman Lukas next introduced President Humberstone, who presented the Section with the Henry F. Nietzel Award for 1954. The award is given annually to the Section which produces the highest numerical gain in membership for the year. For many years, the Membership Committee of the New York Section has been one of the most active in the SOCIETY, working not only for number, but also for quality of membership. It was not surprising then, that the Section should find itself the winner of this award.

In a brief but inspired address, President Humberstone congratulated the Section members for their excellent showing during the past year, and called upon them to do even better for 1954–55.

Mr. Humberstone also commented on the efforts made by other Sections in increasing their respective memberships; he felt sure that the awards were an incentive in that respect.

Acting Technical Chairman H. C. Phelps presented Fairfield N. Stone (M'52), of the J. A. Cunningham Equipment Co., Philadelphia, Pa., speaker for the evening, whose subject was "The Selection and Application of Hard Facing Alloys."

Mr. Stone gave a very comprehensive résumé of the work done to date in this field. He touched upon the

NEW YORK SECTION RECEIVES NIETZEL AWARD

President J. H. Humberstone, *(right)* presents Nietzel Award plaque to Bill Lukas. Secretary J. G. Magrath is on the left.

Al Seiden and Bill Kriewall proudly display award won by the Section for increasing its membership from 481 to 613.

Stan Walter, chairman of the New York Section for 1953–1954, admires plaque won during his term of office.

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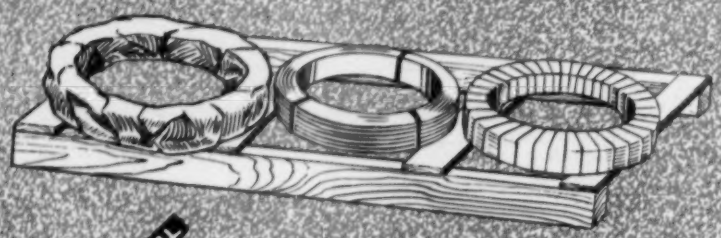

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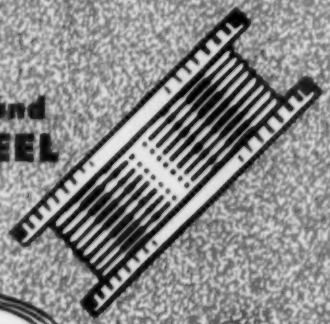

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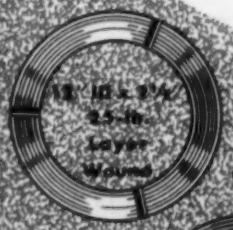

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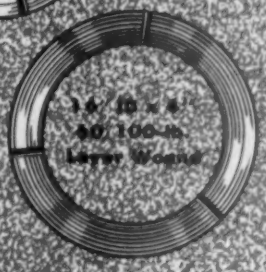

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Armco Gas (GB-45)	Page Armco
Mild Steel (GA-50)	C Gas
Low Alloy (GA & GB-60)	Hi-Tensile M
3½% Nickel (GB-65)	3½% Nickel
.55—.65 Carbon	Medium Carbon
.90—1.10 Carbon	High Carbon
Manganese Bronze	Manganese Bronze
Naval Bronze	Naval Bronze
Stainless Steel	According to AWS analysis

Bare Electrodes

Armco	Page Armco
Low Carbon	Page "CE"
.13—.18 Carbon	Page "B"
.55—.65 Carbon	Medium Carbon
.90—1.10 Carbon	High Carbon

Automatic Welding Wire

Armco	Page Armco
.06 Maximum Carbon	A-S-6
.10 Maximum Carbon	A-S-10
.14 C., 1.75-2.10 MN	A-S-15
.14 C., 1.90 MN., .50 MO	A-S-15-MO
.20 C., 1.20 MN	A-S-20
Medium Carbon	A-S-65
High Carbon	A-S-110
AISI-4130	A-S-4130
AISI-6150	A-S-6150
SAE-9115	NAX-9115
Silicon Killed	Spec. #17
3½% Nickel	3½% Nickel
Stainless Steel Automatic	According to AWS Analysis

Metal Spray Wire

Armco	Page Armco
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.28 Carbon	#25
.40 Carbon	#40
.60 Carbon	#60
.80 Carbon	#80
1.00 Carbon	#100
3½% Nickel	3½% Nickel
Manganese Bronze	Manganese Bronze
Naval Bronze	Naval Bronze
Stainless Steel Metal Spray	According to AWS Analysis

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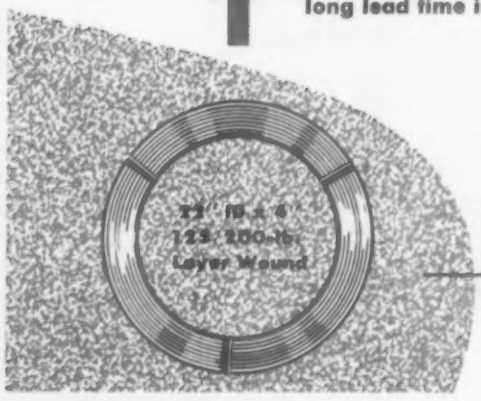

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causes of machinery wear and used an excellent set of slides of various machinery parts, microstructures and tables to give the audience a clear picture of what hard-facing materials to use and how to apply them.

HISTORY OF WELDING

Buffalo, N. Y.—Allan Clarke, of the Ontario Paper Co., was the speaker at the Oct. 28, 1954, meeting of the *Niagara Frontier* Section. Although Mr. Clarke was forced to change his topic at the last minute, he did a very remarkable job in making an impromptu talk concerning the history of welding. An excellent speaker, he held the attention of his audience continuously.

INSPECTION TECHNIQUES

Marion, Ohio—The fourth regular meeting of the 1954–55 season for the *North Central Ohio* Section was held on Dec. 3, 1954, at Dixie's Tea Room.

"Inspection Techniques" was the subject matter of a talk given by Lew Gilbert, executive editor of *Industry and Welding*. Mr. Gilbert's address covered various nondestructive testing methods for welding and weldments. Particular emphasis was placed on the function of Welding Inspectors with respect to welding shop management. Each of the various nondestructive methods was briefly described and applications for the different types were shown. Mr. Gilbert's talk was extremely interesting and very well received.

STORAGE TANKS

Marion, Ohio.—Approximately 50 members and guests attended the January meeting of the *North Central Ohio* Section held on the 7th at the Hotel Harding. The speaker was A. R. Meyer, of Graver Tank & Manufacturing Co. Mr. Meyer related the latest in automatic field welding of storage tanks. He took the audience step by step through the many problems of equipment development and perfection, establishment of procedures and final proving in the field. Horizontal automatic welding is difficult under ideal conditions, but in the field, elevated and to X-ray code, it seemed next to impossible. However, the accomplishment of this proved extremely interesting and informative.

WELDING ALUMINUM

Oak Ridge, Tenn.—Members and guests of the *Northeast Tennessee* Section met on January 17th to hear Charles Bruno, chief welding engineer of the Reynolds Metals Co., give an interesting and informative talk on "Recent Techniques in Welding Aluminum." Among other things Mr. Bruno pointed out that since aluminum alloys possess a thermal conductivity from three to five times greater than ferrous metals, different techniques must be used when welding it. Faster travel speeds must be utilized as must high heat inputs. The use of point contacts in welding jigs is also desirable to compensate for this high thermal conductivity.

The importance of oxide film breakup and removal was stressed. Sound joints, whether soldered or welded, cannot be made unless this film is removed. Recent developments in percussion welding and pipe welding were also discussed. A short question period followed the meeting.

TOUGHNESS IN WELDS

Erie, Pa.—The *Northwestern Pennsylvania* Section met on January 12th at the General Electric Community Center. Technical Speaker was Austin Hiller, manager of the Product Planning and Market Research, Welding Department, General Electric Co. One of the most important steps in determining the engineering value of a structure joined by welding is to evaluate the toughness of its welds. Mr. Hiller's talk stressed the materials

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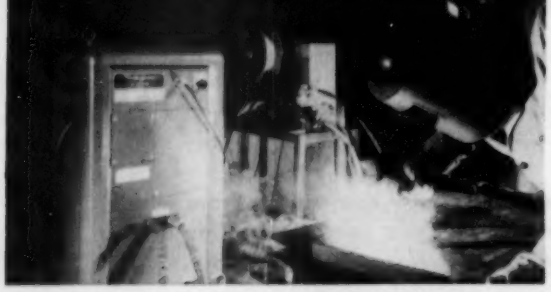

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There's a right Sylvania Tungsten Electrode for every inert gas welding job . . . and a right technique for using the preferred electrode. The following welding helps are the result of field surveys and investigations by the research engineering staff of Rensselaer Polytechnic Institute, and Sylvania Electric Products, Inc.

A. C. WELDING

Metals Welded: Aluminum and magnesium are the metals most frequently welded with alternating current.

Tungsten Electrode: For the usual range of welding applications, Sylvania Zirtung electrodes will give the most dependable and the best performance. Under certain special conditions, Sylvania Thoriated Tungsten and Puretung Electrodes can be used acceptably.

If you have trouble with these problems . . . use these tips . . .

Difficult Starting	Clean work material thoroughly Use superimposed high frequency
Brittle Electrode Tips	Use larger size electrode
Excessive Heat in Electrode or Holder	Use larger size electrode Use centerless ground finish in larger size electrode
Freezing to Workpiece	Try longer arc Use a larger size electrode
Poor Penetration	Use higher current Try helium or a mixture of helium-argon

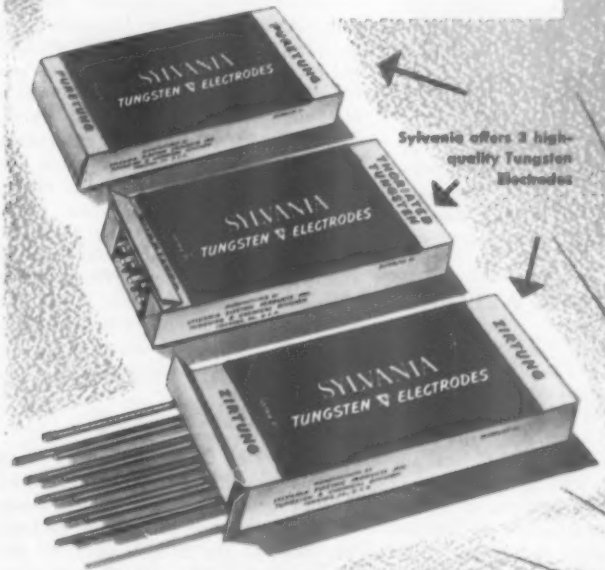
D. C. WELDING

Metals Welded: Stainless steel, inconel, monel and other similar metals.

Tungsten Electrode: For the usual range of welding applications, Sylvania THORIATED TUNGSTEN electrodes will give the most dependable and best performance. Under certain specialized conditions, Sylvania Zirtung and Puretung Electrodes can be used acceptably.

If you have trouble with these problems . . . use these tips . . .

Difficult Starting or Low Voltages	Point the electrode Try touch starting Use argon instead of helium Start fillets with filler rod (if permissible)
Brittle Electrode Tips	Point the electrode Use a larger size electrode
Excessive Heat in Electrode or Holder	Point the electrode Use a larger size electrode Use centerless ground finish in larger size electrode
Metal Pickup, Contamination of Electrode Tip, Erratic Arcs	Use a larger size electrode Use argon instead of helium
Poor Penetration	Use higher current Try Sylvania Zirtung Try helium or helium-argon mixes
Imprecise Welding (especially on fillets or tight corners)	Point the electrode Use starting block
Eroded Electrode	Use a larger size electrode Point the electrode Use a longer arc



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and methods used for obtaining maximum toughness in welds. Mechanical failures of methods in service were also discussed from the design viewpoint.

Following Mr. Hiller's presentation there was shown a film entitled "Resistance Welding of Stainless Steel." This film was shown through the courtesy of the Allegheny Ludlum Steel Corp.

RESISTANCE WELDING

Philadelphia, Pa.—On January 17th the *Philadelphia* Section held a meeting on resistance welding in the Engineers Club. Prior to the meeting, the Executive Committee met with national officers J. G. Magrath, secretary, and F. J. Mooney, assistant secretary, to discuss plans and arrangements for the National Fall Meeting to be held in Philadelphia.

J. R. Wirt, welding engineer of Delco-Remy Division of the General Motors Corp., Anderson, Ind., spoke on the "Fundamentals of Resistance Welding." Mr. Wirt made good use of a number of slides in his most interesting talk showing phases of high production resistance welding of small automotive parts.

MAGNETIC FLUX WELDING

Pittsburgh, Pa.—The regular monthly meeting of the *Pittsburgh* Section for January was held on the 19th in the Auditorium of the Mellon Institute of Industrial Research. A get-together dinner was held in the Hunt Room of the Hotel Webster Hall.

Speaker at the meeting was Howard S. Avery, research metallurgist, American Brake Shoe Co., Mahwah, N. J. "Magnetic Flux Welding" was Mr. Avery's subject. Heat and wear-resistant alloys have engaged much of Mr. Avery's attention and his efforts to define the engineering properties of the important industrial cast heat-resistant alloys are reflected in a series of technical publications. Abrasion testing and the relative performance of many alloys used to resist wear, from austenitic manganese to martensitic irons, have been fields of continual study.

ELECTRODE COATINGS

Portland, Ore.—The fourth meeting of the current season for the *Portland* Section was held on Tuesday evening, January 11th, at the Heathman Hotel. A 30-min social period preceded the dinner. Following dinner, a short business session was held at which time it was announced that the secretary, F. M. Parker, had tendered his resignation, effective immediately; Walt Wyman of Air Reduction Pacific Co., was chosen in his place.

Preceding the speaker's talk, a 20-min color sound film was presented through the courtesy of Air Reduction Pacific Co. Speaker was Harry F. Reid, Jr., of the McKay Co., York, Pa. Mr. Reid presented a talk of very unusual interest entitled "It Isn't Mud." This was probing into the subject of electrode coatings, formulas, etc. A very lively discussion period followed.

SUBMERGED ARC WELDING

Rochester, N. Y.—Bill Weber, welding engineer for the Pfaudler Co., gave a talk on "Submerged Arc Welding" at the Dec. 20, 1954, meeting of the *Rochester* Section. Bill kept his talk at shop level and presented data on joint preparation and actual welding schedules. He had pictures of automatic installations whereby the welder has complete control of turning rolls and welding operations. These installations are costly and are only justified when there is enough volume to warrant their purchase. Many questions were asked and Bill gave direct answers. The paper was well prepared and an instructive evening was had by all.

ELECTRODE COATINGS

Saginaw, Mich.—The *Saginaw Valley* Section held their monthly meeting on January 13th at the High Life Inn in Saginaw. Seventy-five members and guests were present to honor their past chairmen. Si Fisher, the first chairman of the Section, gave a brief review of the history of the Section.

C. R. Miller

The main speaker of the evening was C. R. Miller, Technical Service Division, McKay Co., whose subject was "It Isn't Mud."

In his talk, Mr. Miller covered the general subject of electrode coatings. He broke the function of electrode coatings down into four general areas. These were: control electrical properties, control slag properties, control metallurgical properties, and permitting the use of high-speed manufacturing equipment. He explained with the aid of slides how these properties are obtained by the proper combination of elements in the coating. He classified these elements into four categories (1) binder and extrusion aids, (2) arc stabilizers, (3) fluxes and slag modifiers and (4) alloying materials. Mr. Miller discussed how these elements are combined to obtain electrodes that conform to the various AWS electrode classifications. Mr. Miller closed his talk with a discussion of the steps an electrode takes during manufacture in a typical electrode plant.

PLANT VISIT

Sunnyvale, Calif.—The first plant visitation meeting of the newly formed *Santa Clara Valley* Section, San Jose, Calif., was held on the evening of Dec. 14, 1954, at the 56-acre plant of the Westinghouse Electric Corporation, Sunnyvale, Calif. Following an excellent dinner served in the plant cafeteria, three short talks were given by members of the Westinghouse plant management. George Heiser, division manager, Mechanical Engineering Department, gave the welcoming ad-

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One of 12 groups of Santa Clara Valley Section members and guests who toured the Westinghouse Electric plant at Sunnyvale, Calif.

dress and told of the over-all plant operation. He highlighted the development, engineering and fabrication of the world's largest piece of rotating equipment now in process in the shops. The equipment is the compressor unit for the supersonic wind tunnel now being built at Tallahoma, Tenn. The unit when complete will be approximately 6 stories high and require 216,000 hp attached to the single shaft. Departmental talks referring to specific welding jobs were given by Ed Miller [illegible], and R Moore [illegible], fabrication shop foremen. Following the talks the party was split into 12 groups for the tour through the fabrication shops, main machine shop and assembly floor. The Westinghouse guides were well informed and able to answer any questions as the tour through the plant progressed. Large printed cards were hung on special equipment to further help members and guests know what they were viewing. The 40-ft horizontal boring mill used to handle the extremely large weldments inspired many as to the capacity of the plant. The visitation was voted a real success and, at the same time, highly educational to all in attendance.

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PANEL MEETING

Miami, Fla.—The *South Florida* Section met on January 19th. Six of the members were elected as a panel to answer questions from the floor in the fields of welding, cutting, new processes, a-c vs. d-c machines and safety. The panel members were:

M. C. Stepath, president, Arcair Co., Lancaster, Ohio.
R. D. Robb, representing National Cylinder Gas Co., Miami, Fla.
Prof. John D. Gill, prof. of mechanical engineering, University of Miami, Miami, Fla.
Clyde Streeter, welding superintendent, Miami Shipbuilding Corp., Miami, Fla.
J. C. Shulenberger, director of welding, Eastern Air Lines, Miami, Fla.
A. G. Thompson, vice-president, Miami Welding Supply, Inc., Miami, Fla.

A Stainless Steel and Aluminum Welding Clinic sponsored by the Aluminum Company of America and Crucible Steel Company will be held at the Florida Metals Building in Hialeah on March 23rd, 24th and 25th with some assistance by the South Florida Section.

The Section is running a new membership drive during the next three months. All officers and committee chairmen are competing. Chicken for the winner, hamburgers for the losers; losers pay the bill. Even the chairman is in on this.

PANEL DISCUSSION

Syracuse, N. Y.—The monthly dinner and technical meeting of the *Syracuse* Section was held at the Onondaga Hotel on January 12th. The coffee speaker at the dinner meeting was Elmer Davidson, who is intimately connected with the Crippled Children's Hospitals operated by the Shriners. Mr. Davidson's talk was extremely interesting and outlined in considerable detail the work which is done by the Shrine Crippled Children's Hospitals in the rehabilitation and correction of abnormal bone structures. Mr. Davidson illustrated his talk by a series of before and after pictures showing some of the miraculous work being done.

The technical meeting was featured by a Panel Discussion in which questions covering problems encountered in all phases of welding were answered as fully as possible by one or more of the following group.

Harry Miller, Carrier Corp.—Jigs and Fixtures
Art Raymo, Fitzgibbons Boiler Co., Inc.—Arc Welding
Lowell Hawthorne, Revere Copper & Brass, Inc.—Copper and Brass
Ed Strand, Carrier Corp.—Inert-Gas-Shielded Arc Welding.
Harry Wolfe, General Electric Co.—Resistance Welding
Paul James, San Equipment, Inc.—Submerged Arc

The moderator for the program was A. Cassell of Smith & Caffrey and the judges were Ray Zeh, Lincoln Electric Co., Harry Monson, Welding Engineering & Equipment Co., and Robert Greer of the R. J. Greer Co.

"High jinks," accompanying the decisions of the judges, apart from the more serious phases of the program, were thoroughly enjoyed by the audience.

ARC CUTTING

Sheffield, Ala.—Myron Stepath of the Arcair Co., Lancaster, Ohio, discussed the Arcair process at the January 12th meeting of the *Tri-Cities* Section held at the TVA Auditorium.

INERT ARC WELDING

Tucson, Ariz.—The January meeting of the *Tucson* Section was held at the Tucson Public High School, and was open to the general public.

James Sanderson and Frank Soady, of the Air Reduction Co., presented the program which consisted of a color film on the "Aircomatic Process" and a live demonstration of the subject.

Approximately 250 members and guests saw the film and watched the demonstration; much discussion followed. The attendance exceeded the expectations of the committee and it was necessary to run the film three times in order to give every one an opportunity to see it. Some came from as far as a hundred miles away.

TUCSON OFFICERS

Tucson, Ariz.—The following officers have been elected by the *Tucson* Section for the coming fiscal year:

Chairman—Joe Brennan, 1319 N. Tucson Blvd., Tucson, Ariz.

1st Vice-Chairman—Robert S. Smith, 4740 E. 9th St., Tucson, Ariz.

2nd Vice-Chairman—George Hunt, 1114 E. 7th, Tucson, Ariz.

Secretary—Dean Tillotson, 2606 East Lee, Tucson, Ariz.

Treasurer—Charles Condron, 2348 N. 2nd Ave., Tucson. Ariz.

MOVIE NIGHT

Washington, D. C.—The January 25th meeting of the *Washington* Section was devoted to the showing of a film entitled "Steel Spans the Chesapeake." The film, shown through the courtesy of Bethlehem Steel, showed the erection of one of the world's longest steel highway bridges. Stretching four miles across Chesapeake Bay, the superstructure contains almost every known type of bridge span—beam, deck girder, deck truss, through truss, cantilever and suspension spans. However, little welding is shown since the bridge is mostly of riveted construction.

D-C RECTIFIER WELDERS

Springfield, Mass.—The regular monthly dinner meeting of the *Western Massachusetts* Section was held on January 11th at Blake's Restaurant. Guest coffee speaker was H. Hugo Stahl, No. 1 district director. Mr Stahl gave a comparison of figures on the increase of membership in his district from year to year. He also pointed out the advantages of being a member and what AWS has to offer a member.

National Secretary Joseph G. Magrath presents a copy of the "new" WELDING JOURNAL to the officers of the York-Central Pennsylvania Section at the meeting held on January 13th in York, Pa. Reading from left to right: A. F. Leech, Program Chairman; Bill Delong, Vice-Chairman; R. D. Mann, guest speaker, and Secretary Magrath

Technical speaker was G. K. Willecke AWS, of the Miller Electric Manufacturing Co., Appleton, Wis. Mr. Willecke gave a very interesting talk on the basic principles of the rectifier type D-C Welder in a "down-to-earth" technical discussion, couched in language readily understood by welding engineers and operators not too familiar with complex electrical theory.

The requirements for a direct-current power supply were established and then the various steps in the solution of the problem were analyzed.

The detail function of each component part was thoroughly discussed and examined as to its place in the completed machine.

Special attention was given to the characteristics of the welding current as produced by the various power supplies and the effect of various types of current was discussed in some detail.

Following Mr. Willecke's talk, open discussion was held, which brought out several points of interest.

INERT ARC WELDING

York, Pa.—Robert D. Mann, welding process engineer for the General Electric Co., York, Pa., was the guest speaker at the January 13th meeting of the *York-Central Pennsylvania* Section, held at the West York Inn. Mr. Mann discussed the basic differences between the coated electrode process and the inert-gas-shielded consumable electrode process and the machine requirements to satisfactorily employ them. Slides were used to aid in explaining the development of a welding machine whose characteristics matched those of the inert consumable arc.

Some of the basic problems in the application of this process to the welding of mild steel and the use of carbon dioxide gas as a shielding gas for mild steel welding were discussed.

A 25-min film, consisting of high-speed shots taken at 2500 to 6500 frames per second, showed the arc characteristics and metal transfer of aluminum, copper, magnesium, steel and stainless steel.

A discussion period followed the formal presentation during which many interesting questions were asked with emphasis on mild steel and carbon dioxide shielding gas.

Coffee speaker was National Secretary J. G. Magrath who presented a copy of the "new" WELDING JOURNAL to the Section. Mr. Magrath also reviewed the AWS organization.

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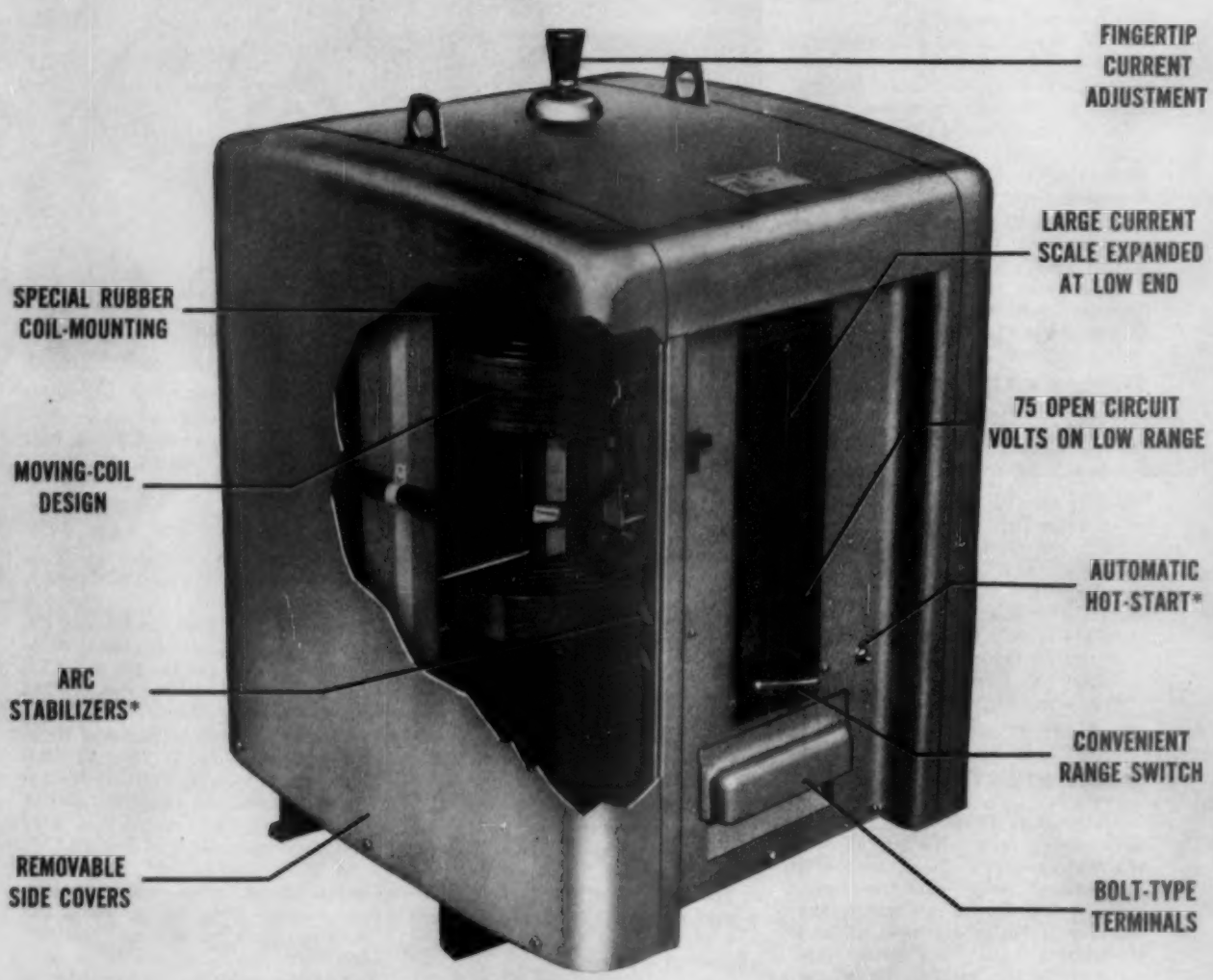

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Arizona: Phoenix—Consolidated Welding Supply

California: Fresno, Los Angeles, Oakland, Sacramento, San Diego, San Francisco, Ventura—Victor Equipment

Colorado: Boulder, Colorado Springs, Denver, Durango, Ft. Collins, Ft. Morgan, Greeley, LaJunta, Longmont, Pueblo, Sterling—Hendrie & Bolthoff

Connecticut: Hartford, New Haven—Harris Company

Florida: Hollywood—Florida Gas & Chemical

Georgia: Athens—Welding Gas Products; Atlanta, Macon—Welding Supply & Service; Augusta—Marks Oxygen; Columbus—Williams Welding Supplies; Gainesville—Welding Gas Products

Idaho: Boise—Gate City Steel

Illinois: Chicago, Moline, Rockford—Machinery & Welder

Indiana: Evansville—Drillmaster Supply; Ft. Wayne, Indianapolis—Sutton-Garten; South Bend—Perry Welding Sales & Service

Iowa: Des Moines—Machinery & Welder

Kansas: Wichita—Standard Products

Kentucky: Louisville—Reliable Welding; Paducah—Henry A. Petter Supply

Louisiana: Alexandria, Shreveport—Hughes Oxygen; New Orleans—Industrial Air Products of the South

Maine: South Portland—Portland Welding Supply

Maryland: Baltimore—Arcway Equipment

Massachusetts: Boston—New England G-E Welding Sales Division

Michigan: Detroit, Saginaw—Welding Sales & Engineering; Grand Rapids—Miller Welding Supply

Minnesota: Duluth—W.P.&R.S. Mars; St. Paul—Production Materials

Mississippi: Greenwood—Delta Oxygen Co.; Jackson—Jackson Welding & Supply

Missouri: Kansas City—Hohenschild Welders Supply; St. Louis—Machinery & Welder

Montana: Billings—Valley Welders Supply; Billings, Bozeman, Cut Bank, Glasgow, Great Falls, Havre, Kalispell, Miles City, Shelby, Sidney, Whitefish—Valley Motor Supply; Butte, Great Falls—Montana Hardware

Nebraska: Lincoln—Lincoln Welding & Supply; Omaha—Baum Iron

New Hampshire: Manchester, Nashua—MacLeod Co.

New Jersey: Elizabeth—Interstate Welding Supply

New Mexico: Albuquerque—Industrial Supply Co.; Hobbs—Western Oxygen; Las Cruces, Silver City—Car Parts Depot, Inc.

NEW G-E 295-amp a-c welder provides big-welder advantages at a small-welder price—only $237**

General Electric's new *295-amp* a-c welder brings you a versatile intermediate rating for: repair work and light production in job shops; industrial maintenance; tack welding and many other general-purpose jobs. It offers many benefits you would ordinarily expect only from a heavy-duty unit:

75 OPEN-CIRCUIT VOLTS—means quick starts, arc stability, and use even with "tough" electrodes like stainless-steel and low-hydrogen.

MOVING-COIL DESIGN—provides pinpoint accuracy in current settings and resettings. Settings can be made in an infinite number of steps over a wide current range.

SPECIAL RUBBER COIL MOUNTING—reduces friction, wear; makes turning of current lever almost effortless.

UNIVERSAL BOLT-TYPE TERMINALS located at the base to eliminate nuisance and hazards of dangling leads.

CONVENIENT RANGE SWITCH permits quick moves from high to low settings without unplugging leads.

For details, ask your G-E welding distributor for bulletin GEC-1305. General Electric Co., Schenectady, N. Y.
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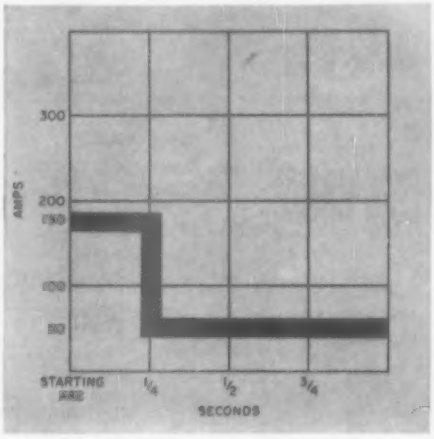
EACH SIDE COVER LIFTS OFF easily, with removal of only 3 screws. And no parts are attached. Simple inside construction can be serviced quickly.



AUTOMATIC HOT-START provides an extra surge of current when electrode touches the work. Arc striking is instant; sticking is reduced. Result: welding speed is increased and weld quality improved.

WIDE CURRENT RANGE spans 25 to 295 amps. Permits wide application. The low end of the range is expanded for fine accuracy.

***Mfrs. suggested resale price. Slight additional charge for freight west of Mississippi and Florida.*

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North Carolina: Charlotte—Dixie Gases; Gastonia—Gastonia Motor Parts

North Dakota: Bismarck, Fargo—Acme Welding Supply

Ohio: Akron, Cincinnati, Cleveland, Columbus, Dayton, Mansfield—Burdett Oxygen; Toledo—Odland Iron Works; Youngstown—Arcway Equipment Co.

Oklahoma: Oklahoma City—Hooper Supply; Tulsa—G-E Welding Sales Division

Oregon: Eugene, Portland—J. E. Haseltine; Medford, Portland—Industrial Air Products

Pennsylvania: Allentown, Philadelphia, Pittsburgh, York—Arcway Equipment

South Carolina: Columbia, Greenville—Welding Gas Products

South Dakota: Deadwood—Hendrie & Bolthoff; Lemmon—Valley Motor Supply Co.

Tennessee: Chattanooga, Knoxville, Nashville—Welding Gas Products; Memphis—Delta Oxygen

Texas: Abilene—M&M Welding Supply; Alice, Corpus Christi—Crane Welding Supply; Alpine, El Paso, Marfa, Pecos—Car Parts Depot; Amarillo—Welding Equipment & Supply; Beaumont—Beaumont Oxygen; Brownsville, Harlingen—Acetylene Oxygen; Dallas—Hill Equipment & Supply; Houston—G-E Welding Sales Division; Lubbock—Welders Supply of Lubbock; Midland—West Texas Welders Supply; Odessa, Pecos—Western Oxygen; Orange—Marine & Petroleum Supply; Pecos—Welding Supply Co.; Plainview—Plains Welding Supply; San Angelo—Southwestern Welding Supply; Texarkana—Hughes Oxygen; Wichita Falls—McGinnis Welding Supply

Utah: Salt Lake City—The Gallagher Co.

Washington: Seattle, Spokane—J. E. Haseltine; Seattle, Spokane, Yakima—Industrial Air Products

West Virginia: Bluefield—Bluefield Supply; Charleston—Virginian Electric; Huntington, Logan—Logan Hardware & Supply

Wisconsin: Milwaukee—Machinery & Welder

Wyoming: Cody, Lovell—Valley Motor Supply Co.

Alaska: Anchorage—Northern Supply

Canada: Toronto—Canadian G.E.

Guam: Agana—Industrial Air Products Div. of Island Equipment Co.

Hawaii: Honolulu—American Factors, Ltd.

New Members

Effective December 31, 1954

MEMBERSHIP CLASSIFICATION

A—Sustaining Member B—Member C—Associate Member D—Student Member E—Honorary Member F—Life Member

ALBUQUERQUE

Bolton, Manuel D. (B)
Henry, Roger E. (C)
Lloyd, David (B)

ARIZONA

Jones, Ray Alfred (C)

BIRMINGHAM

Baldwin, Truman John (C)

BOSTON

Awad, Phillip P. (C)
Balkin, Edward I. (B)
Buckley, J. D. (B)
Dillon, John J. (B)
Peterson, Eric G. (B)

BRIDGEPORT

Fallon, Thomas F. (C)

CHATTANOOGA

Brown, Richard D. (C)
Jones, Tyree E. (C)
Pandelis, C. R. (C)

CHICAGO

Hemzacek, Ray (C)
Kellan, Edwin F. (B)
Kolar, James (C)
Loomis, Robert L. (C)
Puryear, Daniel Oglesby (C)
Oller, Robert E. (C)
Vander Ploeg, Peter A. (B)

CINCINNATI

McKenney, Ronald L. (C)
Wagner, Virgil J. (C)

CLEVELAND

Dolphin, Ben C. (A)
Nickoli, Robert E. (C)
Pestrak, Walter (B)
Pulsifer, Harrison C. (C)
Toffler, Alvin (B)

DALLAS

Corley, Alfred D. (C)
Gray, Joseph J. (C)
Hart, Herman (B)
Hollis, H. D. (B)
Morgan, Howard A. (C)
Van Slyke, Geo. G. (C)
Weiher, Earl R. (B)

DETROIT

Crutcher, Henry B. (B)
Doerer, Richard Paul (C)
Law, James Rothwell (B)
Leupen, Henry C. (B)
Popp, Raymond G. (B)
Teeter, Stuart R. (B)
Varga, Stephen A. (C)

EAST TEXAS

Tait, F. A. (B)

FOX VALLEY

Sappington, C. J. (B)
Sommerfeld, Robert C. (C)

HOUSTON

Fleniken, L. L. (B)
Hill, Philip J. (B)
Pappas, Phillip James (B)
Schilhab, Alvin R. (B)
Schultz, William K. (C)

IOWA

Prough, Dean H. (C)

IOWA-ILLINOIS

Astrosky, Louis (C)
Babnick, A. R. (B)
Carter, M. L. (B)
Curtis, V. D. (C)
Fix, Harry (B)
Goode, Warren (C)
Kemerling, William (C)
Norquist, Ralph (C)
Peterson, Robert E. (C)
Spicer, A. D. (C)
Swett, Ralph (C)
Ufkin, Lawrence (B)

KANSAS CITY

Gassman, Edwin A. (B)
Levitch, Isadore (B)
Montgomery, Robert M. (C)

LONG ISLAND

Bradley, D. Richards II (B)

LOS ANGELES

Bryden, William H. (C)
Carney, C. E. (C)
Kinzie, Freeman X (B)
Lawrence, Paul R. (C)
Schmidt, William, Jr. (B)
Shwitkis, Howard Robert (B)

MADISON

Anderson, Robert C. (D)
Baunt, James G. (D)
Beckman, James (D)
Burke, Robert G. (D)
Carnes, W. R. (B)
Krohg, Arthur J. (C)
Lindberg, Roy A. (C)
Sell, George R. (D)
Smith, Donald E. (D)
Troller, Robert (B)
Vivian, Merrill J. (B)
Whiting, Roger L. (B)

MAHONING VALLEY

Dillon, Homer O. (C)
McNeil, Francis A. (C)
Royer, W. L. (C)

MARYLAND

Jay, Eugene W. (C)
Marks, Robert H., Jr. (C)

MICHIANA

Palfi, Joseph (B)
Wiegand, Ernest C. (C)

MILWAUKEE

Banfay, Stephen (D)
Brown, James W. (C)
Noebel, Carlton E. (C)
Purdy, Kenneth D. (B)
Schmelitsch, Robert C. (D)
Soika, Emil H. (C)
Tucker, Herman (C)
Von Bank, Joseph F. (D)

NEW JERSEY

Bowman, Milford (C)
Lockhart, Bruce Gordon (C)
Minga, Ralph W. (C)
Sieber, Harold H. (C)
Sierra, John A. (C)

NEW ORLEANS

Bayhi, Jas. M. (B)
Bourgeois, Hudson M. (B)
Giardina, George (B)
Elmer, Jackson Forest (B)
Flautt, Robt. A., Jr. (B)
Haddock, G. G. (B)
Janes, F. C. R. (B)
Kennon, Clem R. (B)
Kiehm, Walter J. (B)
Ricaud, Elmo J. (B)

NORTH CENTRAL OHIO

Henderson, C. E. (C)

NEW YORK

Bentley, Thomas R., Sr. (B)
Ebersol, Elmer T., Jr. (B)
Ellis, James George (C)
Ewald, Lux H. (B)
Martin, William Herbert (B)
Massetti, P. A., (C)
Muldore, Joseph (B)
Russell, Leland (B)
Wegge, Thomas (B)

NIAGARA FRONTIER

Densmore, J. G. (B)
Gallinger, C. T. (C)
Heasly, J. R., (C)
Heck, Glenn B. (B)

NORTHERN NEW YORK

Lochner, Roy G. (C)
Lukas, Rudolph J. (B)

NORTHWEST

Maynard, Marcus J. (C)

OKLAHOMA CITY

Hardin, Gerald Wayne (D)

PHILADELPHIA

Heverly, James S. (C)
Pollock, William (B)

PITTSBURGH

Alleman, Eno Joan, Jr. (B)
Bonassi, John (C)
Johnson, A. M. (B)
Krajovic, John R. (B)

PORTLAND

Connor, Eugene W. (B)

RICHMOND

Cobb, W. L. (C)
Grasberger, Charles J. (C)
Grosser, C. E. (B)
Kramer, T. C. (C)
Matthews, W. R. (C)
McIntire, Thomas A. (B)
Smith, Malon F. (C)
Whitlow, J. H. (C)

ROCHESTER

Davies, George H., III (C)

PEORIA

Smith, Howard D. (C)

SAGINAW VALLEY

Cantwell, Charles R., Jr. (D)
Cox, Orville, P. (C)
Lunger, Robert G. (C)
Smith, Paul T. (C)

ST. LOUIS

Heck, Merlyn L. (C)
Kroeger, H. W. (B)
Lehde, James R. (B)

SALT LAKE CITY

Cooper, Clarence B. (C)
May, Samuel Thomas (C)
Sebrands, Kay (B)

SAN FRANCISCO

Jonas, Paul (C)
Leach, Charles M. (B)
Mahoney, R. B. (C)
Marcellin, Wayne J. (D)
Miller, Ralph Kettler (D)
Ozga, Frank (D)
Wilson, Scott P. (B)

SANTA CLARA VALLEY

Archer, S. (C)
Boyajian, E. W. (C)
Canclini, Luke M. (C)
Greer, Otis M. (C)
Havens, John F. (B)
Krankenberg, Verlin L. (B)
Leblanc, Maynard C. (C)
Magallanes, F. (B)
Roderick, Don (B)
Shuirman, Neil (C)
Stroope, George D. (C)
Von Grey, Ed. (C)
Witham, Forest L. (B)

SUSQUEHANNA VALLEY

Atherton, James E. (B)

SOUTH FLORIDA

Bateman, Carl F. (B)
Rosenkrants, Ernst (C)

SYRACUSE

Clough, Harold (C)

TUCSON

Rosenberger, Floyd A. (B)

WESTERN MICHIGAN

Arthur, J. William (B)
Carrigan, W. H. (B)
English, Harold K. (C)
Girton, Preston (C)
Grant, Fred, O. (C)
Krestakos, James (C)
McMahon, Francis Byrne (C)
Simmons, O. B., Jr., (C)
Whalen, Floyd (B)

WICHITA

Hatfield, C. N. (C)

NOT IN SECTIONS

Riquelme, Eugenio, Ing. (B)

Members Reclassified

During the Month of January

NIAGARA FRONTIER

Bevington, Joseph (C to B)

SYRACUSE

Miller, Harry E. (B to A)
Petryck, Leon M. (D to B)

ABSTRACTS OF CURRENT PATENTS

prepared by V. L. Oldham

Printed copies of patents may be obtained for 25¢ from the Commissioner of Patents, Washington 25, D. C.

2,697,954—Method of Cold Pressure Welding—Anthony B. Sowter, Wembley, England, assignor to the General Electric Co. Ltd., London, England.

This patent relates to a method of making a weld along the edges of a pair of ductile metal parts comprising cleaning the areas of contact to be joined to produce pure metal surfaces, placing the parts with the cleaned areas in contacting relation; running the parts past a V-shaped forming pass with the widest part of the V inwardly of the edges so that a force is progressively applied to deform the parts over an area having a maximum width of the order of the thickness of one of the parts. This effects a progressive reduction of the total metal thickness of the parts and causes the metal of the adjoining surfaces at the work area to cold flow into a V-shaped solid phase cold weld joint connecting the parts along the edges.

2,698,548—Method of Pressure Welding—Anthony B. Sowter, Wembley, England, assignor to the General Electric Co. Ltd., London England.

This second Sowter patent is on a method of cold butt welding the ends on wires or the like. Clean end faces of the wires are provided at right angles to the wire axis and such end faces are placed in abutting contact and subjected to endwise pressure axially thereof to force such faces together and upset the metal of the wires equally around the interface and radially of the wires. An extreme interfacial flow of metal radially of the wires at and in the plane of the interface is obtained. The pressure is continued with forces parallel to the axis of the wires to flatten the upset to create a restriction to the radial flow and to intensify the action at the interface. The pressure and compressive action is continued until the upset is a multiple of the cross-sectional dimension of the wire and results is a solid phase bond at the interface.

2,698,813—Process of Silver Brazing Chromium Stainless Steel—Pun Kien Koh, Tarentum, Pa., assignor to Allegheny Ludlum Steel Corp., Brackenridge, Pa., a corporation of Pennsylvania.

This process of brazing chromium stainless steel comprises heating a surface area of the chromium stainless steel to a temperature range from about 1700° F up to the melting temperature of the steel, which heating is performed in an oxidizing atmosphere to effect the intergranular oxide penetration of the surface area. Such surface area is subjected to the action of a fluoride containing material to remove the intergranular oxides from the surface area to provide spaces therein, and silver solder is applied to the treated surface area at a temperature between the melting point of the solder and 1600° F to fill the spaces.

2,698,891—Welding Method and Welding Tool—Amel R. Meyer, Griffith, Ind., assignor to Graver Tank and Mfg. Co., Inc., East Chicago, Ind., a corporation of Delaware.

Meyer's method relates to submerged arc welding for providing a horizontal seam in an upright workpiece surface. The method includes the steps of providing an endless flux belt in rolling contact with the surface below and adjacent to the seam to be provided. A flux is provided on the flux belt and the flux belt is traversed by an arc submerged in the flux and directed along the seam.

2,698,892—Method of Producing Welded Molybdenum—Roy I. Hardin, Chicago, Ill., assignor to Fansteel Metallurgical Corp., North Chicago, Ill., a corporation of New York.

Hardin's method relates to producing fabricated molybdenum characterized by crack-free butt welds. Initially a molybdenum ingot is formed from a mixture of molybdenum powder and a small amount of added carbon to produce from about 0.02% to about 0.09% carbon in the ingot. The ingot is worked to sheet form, and edges of the sheet are positioned in abutting relationship. Such abutted edges then are arc welded in the presence of an inert atmosphere.

2,699,993—Welding Rods for Hard Facing—Samuel G. MacNeill, Kokomo, Ind., and William J. Lindner, Bethlehem, Pa., assignors to Union Carbide and Carbon Corp., a corporation of New York.

A welding rod for hard facing is disclosed in and covered by the patent. The rod is made from an alloy containing: 12 to 16% chromium; 12 to 16% tungsten; 13 to 17% cobalt; 2.5 to 3.5% boron; 0.6 to 1% carbon; up to 1% silicon; up to 0.25% manganese; up to 2% iron; and the remainder of the alloy comprises nickel and incidental impurities.

NEWS of The Industry

International Acetylene Association Convention Scheduled for Houston

The Fifty-Fifth Annual Convention of the International Acetylene Assn. will be held at the Shamrock Hotel in Houston, Tex., on March 22–24, 1955.

A full program of technical papers will feature this meeting. The sessions scheduled are as follows:

Tuesday, March 22nd, 2:30 P.M.
1. Piping Session
2. Marine Session

Wednesday, March 23rd, 9:30 A.M.
Session on "How to Use the Oxy-Acetylene Process Efficiently and Safely"

Wednesday, March 23rd, 2:00 P.M.
1. Oil Field Equipment Session
2. Steel Production and Fabrication Session

The Morehead Medal Dinner will take place on Wednesday evening at 7:00 P.M. It will be highlighted by the formal awarding of the Morehead Medal to Dr. Augustus B. Kinzel, research director, Union Carbide & Carbon Corp. This medal is awarded annually to the person who, at the judgment of the IAA Officers and Board of Directors, has done most to advance the acetylene industry or the art of producing or utilizing calcium carbide or its derivatives.

The Business Meeting, for members only, is scheduled to be held at 9:15 A.M. on Thursday morning, March 24th. The member "Technical Know-How Exchange" will take place on the same morning.

A demonstration-lecture session, starting at 2:30 P.M., will bring the activity to a close on Thursday afternoon.

●

ABS Elects Officers

The 93rd annual meeting of the Board of Managers and the Members of the American Bureau of Shipping was held on January 25th in the Bureau's Board Room, 45 Broad St., New York City.

Walter L. Green, president, presided and expressed to more than 70 managers and members his appreciation of their attendance.

Mr. Green was re-elected president and chairman of the Board of Managers.

David P. Brown was re-elected senior vice-president and technical manager of the Bureau. Jerome B. Crowley was re-elected vice-president in charge of Finance and Administration.

Alfred Blum was re-elected Treasurer and Daniel L. Parry was re-elected Secretary.

The Messrs. Green and Brown are members of the AMERICAN WELDING SOCIETY.

●

Uruguayan Institute of Welding

The Uruguayan Institute of Welding (Instituto Uruguayo de la Soldadura) is the outgrowth of a training course for welders given in 1948 at the Carmelo shipyard. The course was enlarged in scope in 1949 by R. Gasdia and J. Bernardi at the Carmelo Industrial School, a branch of the Technical University of Uruguay. In 1952 the instruction was transferred to the Faculty of Engineering and Surveying in Montevideo. The courses in welding, including procedures, metallurgy and electricity, were so well attended that an Institute of Welding was organized in December 1952.

The new Institute consists of the general membership, which elects a President and 16 members of the General Council. The President heads the General Council and the Executive Committee. The General Council decides on the over-all scope of the Institute's activities. The Executive Committee contains four members from the General Council and two Chairmen of Technical Committees. It is the administrative body of the Institute. There are four technical committees: (1) Advisory Committee on Technical Matters, (2) Standards Committee, (3) Publication Committee and (4) Professional Committee which also serves to promote membership.

●

Stud Manufacturing Operations

The Nelson Stud Welding Division of Gregory Industries, Inc., has announced the resumption of limited stud manufacturing operations at its western regional headquarters in San Leandro, Calif.

The manufacturing operations will be located at the factory branch warehouse at 440 Peralta Ave., San Leandro, the original location of the Nelson stud welding business which moved to Loraine Ohio, in 1945.

Ampco-Weld Distributors

Several distributors have been established to handle the Ampco-Weld line of electrodes and accessories according to the announcements released by the Resistance Welding Division of Ampco Metal, Inc., Milwaukee, Wis.

The new distributors include Brady Supply Corp., Elmira, N. Y.; C. D. Genter Co., Chattanooga, Tenn.; Keen Machinery Co., Cincinnati, Ohio; Walter Pestrák, Cortland, Ohio, and Georgia Supply Co., Jacksonville, Fla.

●

Machine Tool Design Award Program

Dr. E. E. Dreese, chairman of the Board of Trustees of The James F. Lincoln Arc Welding Foundation, Cleveland, Ohio, has announced the sponsorship by the Foundation of a $12,000 Machine Tool Design Award Program. Awards will be made for papers describing the use of arc welding in machine tool design.

This program is open to all persons who are or have been engaged in the design or making of metal-cutting or metal-forming machine tools. A total of 15 awards will be made, with a top award of $3000 and other main awards of $2500, $2000, $1500, $1000 and 10 awards of $200 each. Awards will be made to the 15 best papers submitted to the Foundation describing the use of arc welding in the design of either a component part or complete machine tool. The tool may be either a metal-cutting or metal-forming tool as defined by the National Machine Tool Builders' Assn. The work or equipment described must have been planned or executed in the period from Jan. 1, 1953 to July 15, 1955. The design may be one that either has been manufactured or one that is planned to be manufactured. It may be either a redesign or a new design. The program closes July 15, 1955.

The award program is open to all persons in the United States and its possessions who have actually been engaged in the design or manufacture of the equipment described. A single author or a group of authors may submit a paper. A booklet describing the program and how to participate is available from the James F. Lincoln Arc Welding Foundation, Cleveland 17, Ohio.

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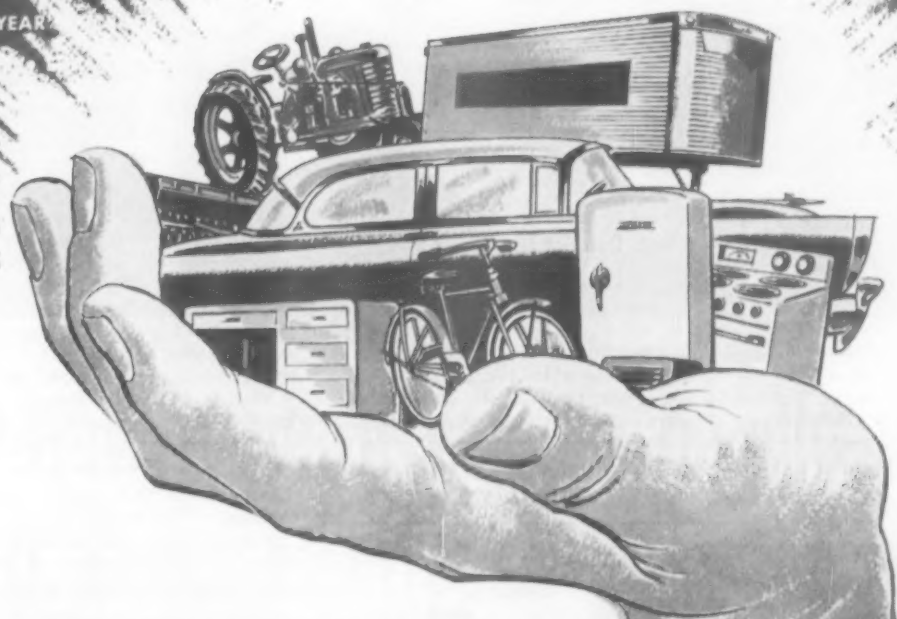

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Welding Engineering Scholarship

Award of the 1954–55 Welding Engineering Scholarship of $150 to William F. Brown, a senior in Ohio State University's College of Engineering, has been announced today by Dean Gordon B. Carson.

The Welding Engineering Scholarship Fund was established in 1942 by an anonymous donor and supports one senior scholarship. In addition to scholarship, a candidate's scholastic record and possibilities for success in professional work are considered in making the award.

The department of welding engineering at Ohio State is one of two degree-granting departments in that field in the world. The other department is at Osaka University in Japan.

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Welding Instructions on Records

"Eutectic Welding Institute on Records," a 2-hr 33-rpm album of the week-long courses held at Eutectic Welding Alloys Corp., Flushing 58, N. Y., is now available to business.

Along with the unbreakable records, Eutectic furnishes a 60-page manual with "blackboard" diagrams, a reference copy of the record script, a series of 8 tests with answers and a handy indicator showing subjects covered on various parts of the records.

Welding procedures such as torch manipulation, electrode angle and travel application heats, the use of fluxes and selection of correct welding material are given. "Low-heat input" joining of aluminum, cast iron, copper and copper alloys, steels, stainless steels and other base metals is covered. Numerous industrial applications for welding, as well as procedures for heating methods, are also described.

The records are free on a loan basis when available from Eutectic District Engineers.

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Airco District Manager

L. O. Geiger, manager of the Dayton, Ohio, district for Air Reduction Sales Co., retired Jan. 1, 1955, it has been announced by J. H. Keeney, regional sales manager in the central region. Mr. Geiger was with Air Reduction 38 years and supervised the sales activities in the Dayton area since 1927 when he was appointed to the post of district manager.

C. R. Grange, assistant district manager, succeeds Mr. Geiger. Mr. Grange has been with the company since 1934 when he started as a student salesman. In succeeding years he held the position of salesman in the Jersey City, N. J. and Philadelphia, Pa., district offices. He served in the Navy from 1942 to 1945, returning to Air Reduction as assistant sales manager in the St. Louis district. Mr. Grange has been in the Dayton district since June 1954.

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National Electric Welding Elevates Key Employees

Following the Annual Meeting of the Stockholders of National Electric Welding Machines Co. of Bay City, Mich., held on Jan. 10, 1955, the newly elected Board of Directors named the following officers for the ensuing year: Edward C. Smith, president; William L. Mueller and Howard C. Cogan, vice-presidents; Charles E. Shearer, secretary-treasurer, and Norman A. Witzleben, assistant secretary-treasurer.

The Board, on the recommendation of Mr. Smith, appointed Simon Fisher as the new general manager and Dean L. Knight, newly elected member of the Board of Directors, as the new plant manager. Mr. Smith, former general manager, and Mr. Mueller, former plant manager, continue as president and vice-president, respectively.

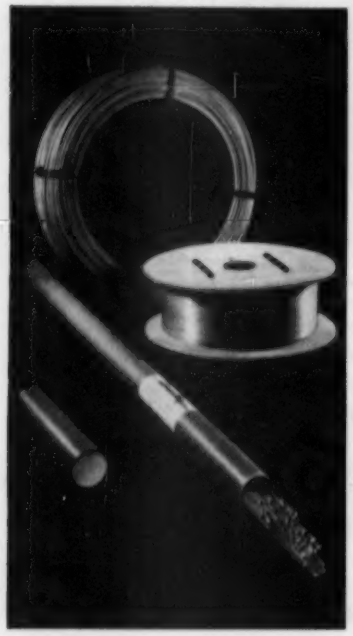
•

Permagile Corp. Formed

Formation of a new company, Permagile Corporation of America, 300 Fourth Ave., New York City, to manufacture and sell new type plastic resin compounds for sealing and welding many dissimilar nonmetallic materials, was announced recently by Dr. J. A. Neumann, president of the American Agile Corp., Cleveland, who will likewise head the new firm.

Officers of the new concern, in addition to Dr. Neumann, are Theodore I. Leston, former vice-president in charge of research, development and production for the Eutectic Welding Alloys Corp., New York, who will serve as executive vice-president of Permagile; William B. Kriewall, former chief engineer of Eutectic, who will serve as vice-president in charge of production and development for the new firm; and G. Vajda, who is also president of G. Vajda & Company, Inc., New York and Tokyo.

The Messrs. Neumann, Leston and Kriewall are members of the AMERICAN WELDING SOCIETY.



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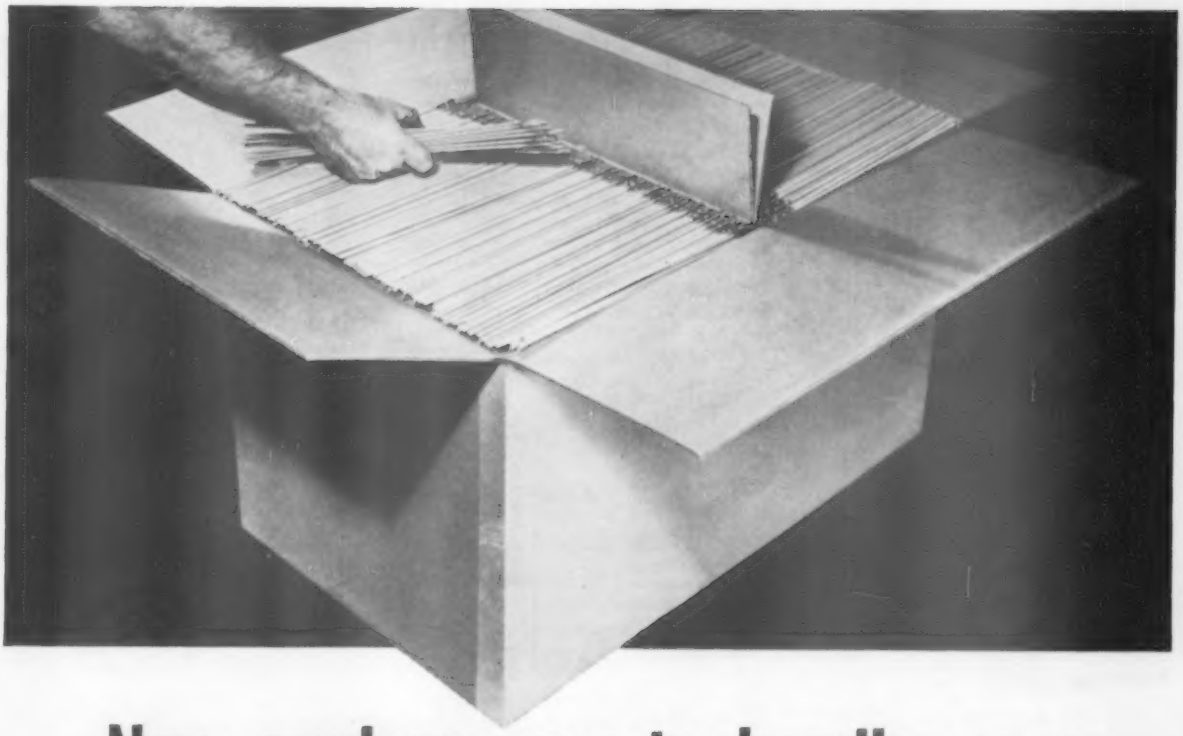

New modern way to handle your A. O. Smith electrodes

Production-Pak containers work for you 'til the last rod is used

If you use electrodes in *quantity*, this is for you! A. O. Smith's new Production-Pak containers start paying off the moment they hit your plant.

● **HANDLING EASE.** A single Production-Pak contains approximately 2,000 lbs. of electrodes . . . as many electrodes as you'd get in 40 ordinary boxes. Think of the time and labor saved in unloading alone. Think of the freedom from costly palletizing and de-palletizing.

● **STOREKEEPING CONVENIENCE.** Production-Pak is its own storage bin. Electrode distribution is faster, easier. No need to break open individual smaller boxes. Just pass out the rods by the handful.

● **INVENTORY SIMPLIFIED.** Fewer boxes to count. Inventories stay better-balanced . . . because electrodes are easier to dispense on a first-in, first-out basis.

Ask your A. O. Smith distributor about Production-Pak today . . . or write A. O. Smith Corporation, Welding Products Division, Milwaukee 1, Wisconsin.

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WELDING PRODUCTS DIVISION
Milwaukee 1, Wisconsin
INTERNATIONAL DIVISION: MILWAUKEE 1

Airco Paper Wins Award

The American Institute of Electrical Engineers has awarded second prize in the Industrial Division to the paper entitled, "Atomic Film Cathode in Inert-Gas-Shielded Metal Arcs," co-authored by W. J. Greene, Assistant Director of Metallurgical Research, and Dr. A. Muller, Director of Metallurgical Research, both located in the Murray Hill, N. J., Research Laboratories of Air Reduction Co., Inc.

The paper details Air Reduction's development work in the field of emissive coated wires for the inert-gas-shielded metal-arc welding process. The new coating, which is reported to form an atomic film over the molten end of the electrode, is expected to extend the utility of the inert-gas-shielded metal-arc welding process, particularly in the field of alternating current and straight polarity welding.

•

Resistance Welder Representatives

Two new appointments have been announced by Boyd Gresey, general manager, Resistance Welder Corp. Wayne L. Cummins will be district representative for the Dayton-Cincinnati area, and L. E. Swanson will be district representative for the St. Paul-Minneapolis area.

Wayne L. Cummins

L. E. Swanson

Mr. Cummins is an alumnus of General Motors Institute, served over four years in the Army Air Forces and has been active in resistance welding for the past seven years.

Mr. Swanson has been actively engaged in the sale and servicing of resistance welders and controls for the past five years. His earlier work included designing of electrical distribution and transmission systems.

Mr. Swanson is an alumnus of the University of Minnesota and he is a registered electrical engineer.

Resistance Welder Corp. designs and builds resistance welding equipment, allied fixtures, tooling and automations.

•

Delta Welder Corp. Organized

The formation of the Delta Welder Corp. in Detroit was recently announced by Harry E. Day, president.

In outlining the objectives of the Delta Welder Corp., Mr. Day stated that his organization is singularly qualified in the design, engineering, and building of completely automatic multiple point welding machines. Further, he pointed out, the Delta Welder Corp. will engineer, design and manufacture high production special and standard resistance welding machinery and automation equipment for the metal fabricating industries.

Harry E. Day

Delta Welder Corp. plant and executive offices are located at 8525 Livernois, Detroit 4, Mich.

•

Alloy Rods Creates New Posts

E. J. Brady, president of the York, Pa., electrode firm, announced recently the establishment of two new posts within the organization with the following new appointments:

George M. Hohmann

George M. Hohmann, who for the past year has been manager of the company's Pacific Coast Division, has been named vice-president in charge of production for both plants.

E. R. Walsh III has been appointed to the post of vice-president in charge of sales, and will be responsible for all company marketing, sales and promotional activities. Mr. Walsh has served in the capacity of general sales manager for the past three years.

E. R. Walsh, III

•

New GE Structure

Completion of a new structure to house General Electric's welding sales branch at Tulsa, Okla., was announced recently by E. K. How, sales manager for the company's Welding Department.

The building, which includes offices, service and warehouse facilities, will serve Colorado, Wyoming, Kansas, Arkansas and Oklahoma. According to Mr. How, the new G-E facility is designed to improve service to the pipeline industry in the Southwest, as well as to other users of welding equipment, electrodes and accessories.

A. O. Smith developed the exclusive techniques and machines for making non-corrosive "lined" pressure vessels for use in oil refineries, pulp digesters and chemical plants. The non-corrosive liners are spot welded to the inside of the vessel with thousands of welds.

Square D Welder Control proved itself in production by increasing speed, adding to the strength of the welds and reducing repairs

SQUARE D *Electronic* WELDER CONTROL

provided these improvements on one hard-to-weld alloy:

SPEED
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PERSONNEL

Pellini Wins Award

The Board of Managers of the Washington Academy of Sciences has selected W. S. Pellini to receive the 1954 Award for Scientific Achievement in the Engineering Sciences.

The award was granted "in recognition of Mr. Pellini's notable contributions in the field of metal processing."

The award is based on Mr. Pellini's work in welding and casting metallurgy prior to his recent promotion to Superintendent of the Metallurgy Division at the Naval Research Laboratory, Washington, D. C. He is a member of the AMERICAN WELDING SOCIETY.

●

Welding Engineering Instructor

Return of William L. Green as an instructor to the staff of the department of welding engineering at Ohio State University has been announced by Dean Gordon B. Carson. Holder of two degrees from Ohio State, Mr. Green had served as an instructor in the department prior to taking a leave-of-absence to serve a tour of duty with the U. S. Navy Air Forces, Atlantic Fleet, since February 1953. He will be doing teaching and research in welding design and arc metal transfer in the department.

●

Hobart Representative

Robert E. Roediger has just been appointed Hobart Brothers Sales and Service Representative for the Birmingham Alabama Area. He will make his headquarters at 2657 Park Lane Court, East, under the name of Hobart Welding Supply Co. Mr. Roediger will have full charge of servicing and distribution of a complete line of modern arc welding equipment, supplies and electrodes.

Robert E. Roediger

Robert M. Wilson, Jr.

Wilson Assigned to Power Field

The assignment of Robert M. Wilson, Jr., to act as a development engineer in the power field for the Development and Research Division of the International Nickel Co., Inc., has been announced by F. L. LaQue, vice-president and manager of the division.

In announcing Mr. Wilson's appointment, Mr. LaQue said that this action recognizes the growing importance of the power field as a consumer of nickel alloys and the greater needs of the industry for better materials to permit operations at the higher temperatures and pressures which characterize modern power plants.

A graduate in mechanical engineering of Cornell University, Mr. Wilson has been with International Nickel since February 1946, when he joined the company as an engineer in the Technical Service Section of the Development and Research Division. Previously, he had been welding engineer at the Bloomfield, N. J., works of the General Electric Co.

Mr. Wilson is a member of the AMERICAN WELDING SOCIETY and the American Society of Mechanical Engineers, and has been active in Boiler Code work.

●

Custance Promoted

Donald E. Custance has been appointed assistant Sales Manager by Progressive Welder Sales Co. He has been associated with this company for over 15 years in various engineering and sales capacities.

McMaster Joins Ohio State

On Jan. 1, 1955, Dr. Robert C. McMaster assumed his duties as Professor in the Department of Welding Engineering at Ohio State University. Previously, he had been a member of the staff at Battelle Memorial Institute in Columbus, Ohio. Dr. McMaster has taught at Case Institute of Technology and California Institute of Technology, and has been associated with the research laboratory of the General Electric Co. and the Naval Ordnance Laboratory in Washington, D. C. During World War II, Dr. McMaster supervised research in welding and X-ray nondestructive testing of aircraft materials.

Dr. Robert C. McMaster

In his new academic role, Dr. McMaster will be able to pursue his specialties of resistance welding, nondestructive testing, electronic welding and arc phenomena through the channels of both teaching and research. He holds active membership in a number of societies, including the AMERICAN WELDING SOCIETY. He is author or coauthor of 62 published technical papers.

OBITUARY

Ben H. Allen

Ben H. Allen, superintendent of the Plate Shop, Mosher Steel Co., Houston, died suddenly Jan. 18, 1955, in a Houston hospital, at the age of 49.

Mr. Allen was one of the most active workers in the AMERICAN WELDING SOCIETY, having served either as an officer or a director of the Houston Section for many years. He was a charter member of the Houston Section.

He was Chairman of the Houston Section in 1952–53 at the time of the first annual Welding Society Show which occurred at the Shamrock Hotel, Houston. He was a real sparkplug in helping to organize and lead this first AWS Show to its success.

Mr. Allen contributed greatly to the growth of the Houston Section for the past few years. It was during his tenure as chairman that the Houston Section was winner for the third consecutive year of the Henry Nietzel award for greatest numerical net gain in membership.

Ben was extremely interested in our youth and was very active in promoting our AWS student membership. His outstanding influence on our youth was probably best evidenced as follows: he first became affiliated with Boy Scouts when he joined this organization on his 12th birthday, becoming an Eagle Scout in 1923, and Scoutmaster of the First Christian Church Troop in 1933. In 1935 he became an adult worker in the Heights District of the Sam Houston Area Council. In 1952 the Council recognized his efforts with the Silver Beaver award, the highest honor for volunteer adult workers.

Mr. Allen was a Rice Institute alumnus, a member of St. Mark's Methodist Church, and the Woodman Masonic Lodge 1157.

Ben Allen's enthusiasm, initiative, drive, and pleasant, optimistic attitude will be missed by all his associates.

Robert G. Boss

Robert G. Boss, 35, of Rochester, N. Y., died at his home January 11th, after several months' illness. Bob served from 1951 to 1953 as a member of the Executive Committee of the Rochester Section AWS, and will be remembered by his fellow associates for his friendliness and his even temperament.

After serving three years in the U. S. Air Force, he was employed by Jackson Welding Supply, William A. Curtice Corp., and as regional manager of All-State Welding Alloys Corp. He was also a member of the American Engineering Society.

He is survived by his wife, Ruth; a son, David; his parents, Mr. and Mrs. Louis Boss; two sisters and three brothers.

A. Ralph Ellis

A. Ralph Ellis, 73, chairman of the board of directors of Pittsburgh Testing Laboratory, died Dec. 24, 1954, in the Pittsburgh Hospital.

Mr. Ellis, who was born in old Allegheny, lived at 6963 Edgerton Ave., Point Breeze. He was associated with the firm he headed throughout his professional career.

After graduating from Cornell University in 1906 as a civil engineer, Mr. Ellis joined Pittsburgh Testing Laboratory as a laboratory worker. In succession he became chief engineer, manager of the New York branch, general manager and director, vice-president, president and chairman.

Mr. Ellis was a registered professional engineer in eight states. He was a member of several societies one of which was the AMERICAN WELDING SOCIETY. He was a long-time, interested and active member of the Pittsburgh Section.

Mr. Ellis is survived by his wife, Mrs. Miriam W. Ellis, and a son, A. Ralph Ellis, Jr.

Attend

National Spring Meeting

June 7-10

Kansas City, Mo.

EMPLOYMENT SERVICE BULLETIN

Position Vacant

V-318. Sales Service Engineer. Specialty steel producer desires a young man with engineering background to sell and service on stainless, hard-facing and tool steel electrodes. Midwest territory operating out of Chicago. State qualifications and salary desired.

Services Available

A-663. Welding Engineer with practical and formal education desires responsible position in sales, engineering, or production, in that respective order. Background: M.E. degree—trade school graduate—apprenticeship plus 7 years' practical experience—5 years' engineering and sales representative in development and manufacturing problems. Experienced with inert-arc processes, automatic and manual, all oxy-acetylene processes, electric welding, ferrous and nonferrous metals. Age 35. Married. All replies answered.

A-664. Welding Engineer. B.S. in Metallurgy. Thirteen years' diversified metallurgical work largely in the aircraft industry. Broad experience in sheet metal fabrication for jet applications. Specialist in high temperature brazing and welding of high temperature alloys. Ability to set up laboratory development projects and handle production problems. Author of papers. Desires responsible position with advancement possibilities.

A-665. Welding Supervisor, 34 years of age, desires position in supervision, engineering or sales. Has broad background and knowledge of welding, and the allied metalworking fields. His experience has covered supervision, engineering, sales, production, electrical equipment installation and maintenance; hard-surfacing applications and aircraft and job shop welding. Résumé upon request.

BOUND VOLUMES OF 1954 JOURNAL NOW AVAILABLE

Bound Volumes of THE WELDING JOURNAL for the year 1954 are available in black imitation leather covers, together with a comprehensive subject and authors index. Price $15, including postage.

This volume, comprising a total of 1238 pages in the JOURNAL and an additional 624 pages in the *Welding Research Supplement*, represents a veritable encyclopedia of information in the welding field. Copies may be ordered through the AMERICAN WELDING SOCIETY, 33 W. 39th St., New York 18, N. Y.

NEW LITERATURE

AWS Metallizing Standard

Heat corrosion, a subject of much concern in the oil refinery, chemical plant, furnace industry and other high temperature fields, is the subject of the third report in the series under the general title, "Recommended Practices for Metallizing." This Standard, "Part 1C—Application of Metallized Coatings to Protect Against Heat Corrosion," provides an authoritative source of information dealing with the application of metallizing for providing protection against heat corrosion for temperatures up to and over 1800° F.

Equipment requirements and methods of surface preparation are listed in detail. The metallizing procedure, including wire and sealer compositions, and inspection tests are also given as well as typical coatings for specific applications.

Copies are available at 50 cents each from the AMERICAN WELDING SOCIETY, 33 W. 39th St., New York 18, N. Y.

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Arc Welding News

Volume XI, No. 4, "Hobart Arc Welding News," a 16-page booklet of interesting photographs and articles on welding from all over North America, is now available. Copies are mailed free of charge to anyone interested in welding. To get your copy, write the Hobart Brothers Co., Troy, Ohio.

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Brazing Filler Metals

A tabulation of "Brazing Filler Metals for Electron Tubes" has been prepared by Dr. Walter H. Kohl, Electronics Consultant on Materials and Techniques, P. O. Box 426, Los Altos, Calif., and is available as a wall chart (21 x 24 in.) at the following rates: 1 to 5 at $3.00 each, 6 to 12 at $2.75 each and 13 to 25 $2.50 each. Solidus and liquidus temperatures are given both in degrees centigrade and fahrenheit and percentage compositions of alloys are listed. As the title of the chart suggests the 58 listed metals and alloys are restricted to low vapor pressure materials. Trade names are given only for two entries but blank columns permit the user to enter his preferred materials by the trade name matching the listed alloy composition and their supplier.

Welding Research Abroad

The Welding Research Council, 29 W. 39th St., New York 18, N. Y., has just created a new publication entitled "Welding Research Abroad" as a service to its subscribers and research workers. This publication, the first issue of which consists of 32 pages, is in photostat form. The first issue includes a considerable amount of background material relating to the activities of the International Institute of Welding. The publication, which will be issued quarterly, will include timely research articles on specific subjects, reports of the different Commissions of the IIW, translations or abstracts of important foreign articles and items of a similar nature.

•

Filler Metal Chart

The Arcos Corp., Philadelphia, Pa., has issued a new colored chart keying by use its full line of filler metals for welding. Each rod and electrode is identified by the metal for which it is suited, forms available and methods with which it is used. Methods covered are metal arc, submerged arc, tungsten electrode, metallic electrode, oxy-acetylene and atomic hydrogen. Write directly to company for copy.

•

Projects in Metal

The Lincoln Foundation is making available, at small cost, an 80-page booklet entitled "Projects in Metal for Home and School Shop."

It has been reprinted from a larger Foundation book, *Farm Arc Welding*, but the projects described are useful for any home regardless of whether it is in the country or in the city. Projects include such things as: tree guard, stair railing, furnace, clothesline post, lawn furniture, wrought iron work, shop tools, air compressor, pipe bender, vise, ladders, stools, child's slide, swing, boat trailer, grill, and toy tractor.

It contains 122 drawings and photographs and is bound in heavy paper. Price 50¢ per copy in the United States, postage prepaid; 75¢ per copy elsewhere. For your copy, write to the James F. Lincoln Arc Welding Foundation, Cleveland 17, Ohio.

Ampco Welding News

Ampco Metal, Inc., has released the fourth-quarter issue of its publication, *Ampco Welding News*.

The feature article in this issue explains how Pines Bender equipped with Ampco wiper dies and Ampco-Trode overlaid mandrels have saved the aircraft industry as much as $14,000 per plane.

Free copies of this issue can be obtained by writing Ampco Metal, Inc., 1745 S. 38th St., Milwaukee 46, Wis.

•

Safety Equipment

The 1955 issue of *Everything in Safety*, a catalog of personal protective equipment and industrial safety devices, has just been released by the General Scientific Equipment Co., 2700 W. Huntingdon St., Philadelphia 32, Pa.

This new catalog covers respiratory devices, eye protection, hats, gloves, carboy pumps, drum pumps and miscellaneous industrial safety equipment and many unusual safety specialties for use in industries, mines, utilities, and farms.

It contains 130 pages, fully illustrated. Copy sent free on request to the company.

•

Airco Facts

The Air Reduction Co., 60 E. 42nd St., New York 17, N. Y., has recently published "Facts About Air Reduction" which describes generally the products and services of the various divisions of the company. Copies available upon request.

•

Stainless Tubing Fabrication

A new 36-page catalog on fabricating and working stainless tubing and pipe has been issued by the Alloy Tube Division of the Carpenter Steel Co., Union, N. J. Each of 11 types of fabrication, including welding, is described separately, with drawings and photographs augmenting the information. Helpful hints tell how to design for economical fabrication. A basic section gives the fabricating characteristics of ferritic, austenitic and certain specialty steels. Write to above address for copy of catalog.

Resistance Welding

Sciaky Bros., Inc., 4915 W. 67th St., Chicago, Ill., Vol. 4, No. 3, *Resistance Welding at Work*, covers several case histories of Sciaky patented three-phase electric-resistance welding equipment applications.

To receive your copy, write direct to the Chicago plant.

•

Copper Alloy Specifications

Publication of a tabbed, easy-to-use, 24-page reference manual, containing the latest copper and copper-alloy specifications, is announced by the American Brass Co., an Anaconda subsidiary. Included in this completely revised edition are ASTM, ASME, AWS, SAE, AMS, Federal, Military, Army, Navy, and Joint Army-Navy specifications.

Copies available free, on request. Write for Anaconda Publication B-34R; American Brass Co., Waterbury 20, Conn.

•

CO_2 Applications

"CO_2: Applications Unlimited," a colorful, new booklet describing the many uses to which carbon dioxide is now being put in industry, has been published by The Liquid Carbonic Corporation. Free copies are available upon request to: The Liquid Carbonic Corp., Advertising Department, 3100 S. Kedzie Ave., Chicago 23, Ill.

•

Brazing News

Handy & Harman, 82 Fulton St., New York 38, N. Y., has just released their latest copy of "Low-Temperature Brazing News." Ask for Bulletin No. 67.

•

Stainless Steel Chart

The Stainless Steel Wall Chart, printed by Alloy Metal Wire, presents the complete engineering properties of stainless steel wire, rod and strip. When fully opened, the Stainless Chart measures 17 x 22 in., and folds neatly into an $8^1/_2$ x 11-in. folder.

Over 20 different stainless steel alloys in the martensitic, ferritic and austenitic groups are described, and their properties listed. The chart is printed on heavy, durable paper with a dull finish to eliminate glare.

Charts may be obtained by writing to E. H. Mann, Alloy Metal Wire Division, H. K. Porter Co., Inc., Prospect Park, Pa.

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automatic RBM manipulator...
KING SIZE WELDING HEAD PILOT
(STANDARD AND SPECIAL)
—power car moves up or down track at variable welding speeds (available on fixed base)
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6' 6'
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See you at Booth 245, Kansas City Welding Show June 8-10

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AVAILABLE FOR QUICK DELIVERY
SEMI-AUTOMATIC MANIPULATOR... MOUNT SEMI-AUTOMATIC CONTROL CABINET ON SAM...CONVERT THE HEAD INTO A SPEEDY SELF-PROPELLED TOOL:
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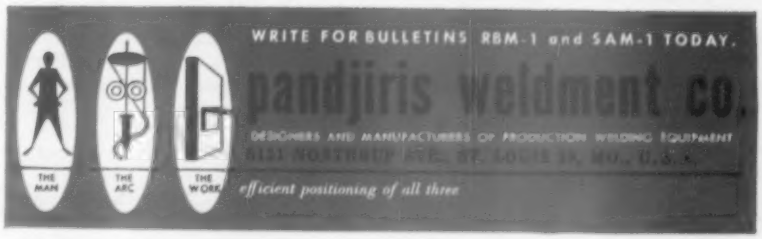
WRITE FOR BULLETINS RBM-1 and SAM-1 TODAY.
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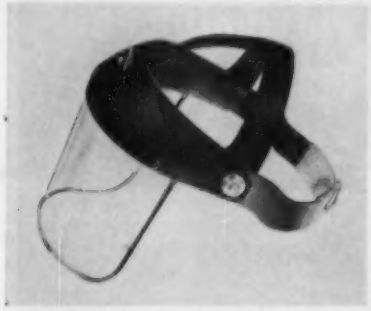
DESIGNERS AND MANUFACTURERS OF PRODUCTION WELDING EQUIPMENT
THE MAN
THE ARC
THE WORK
efficient positioning of all three

NEW PRODUCTS

Face Shield

A new face shield has been designed and is now being manufactured by the Fibre-Metal Products Co. Trade-named "Monarch," it reportedly provides the utmost in protection against low impact exposures. The shield pictured, one of four similarly designed, has



a spark deflector and allows for the quick interchangeability of a general-purpose window, a window for heat protection or a special window for acetylene welding or scarfing.

Further information may be obtained by writing the Fibre-Metals Products Co., Chester, Pa.

●

Rectifier Welder

A new 300-amp d-c selenium rectifier type arc welder with remote control is being announced by Hobart Brothers Co., Troy, Ohio.

The 5-range control switch of this welder provides coarse adjustment with generous overlap between ranges to provide dual control. The rheostat, which can be removed for remote control, provides continuous control by adjusting the direct current through the control coils. These controls provide a wide welding current range of 35 to 425 amp for use with a variety of electrode sizes.

For complete information write Hobart Brothers Co., Troy, Ohio.

●

Control for Small Spot Welders

On page 1208 of the December 1954 issue of THE WELDING JOURNAL, there appeared a brief description of an electronic welder control for use with manually operated small spot welders; however, the name of the manufacturer was not given.

For further information concerning this control, please write directly to Taylor-Winfield Corp., 1052 Mahoning Ave., N.W., Warren, Ohio.

●

Temperature Indicators

The Tempil° Corp., 132 W. 42nd St., New York 11, N. Y., announces that the company can now supply from stock Tempil° pellets for 425 and 475° F.

The inclusion of these two intermediate temperature ratings makes Tempil° pellets available in $12^1/_2$-deg steps from 113 to 400° F, in 25° F intervals from 400 to 550° F, and 50-deg steps from 550 to 2500° F. They are packaged in tubes containing 20 Tempil° pellets of a single temperature rating, each pellet indicating its specified rating with an accuracy of ±1%.

●

Phos-Trode C

Ampco research laboratories announce the release of an improved Phos-Trode electrode which has a phosphor bronze Grade C (AWS-ASTM E Cu Sn C) core and an entirely "new," patented covering. Phos-Trode is a heavy covered, shielded-arc electrode for the metal-arc welding of copper, tin bronzes, brasses, galvanized iron, cast and malleable iron and dissimilar metals in all positions.

Send to Ampco Metal, Inc., 1745 S. 38th St., Milwaukee 46, Wis., for a copy of Bulletin 54L-150 describing Phos-Trode in detail.

Cast-Iron Electrode

A low-cost, nonmachinable, cast-iron electrode, primarily for repair and reclamation where finish may be ground or left as welded, is announced by All-State Welding Alloys Co., Inc., 249–55 Ferris Ave., White Plains, N. Y. Designated as All-State No. 6 High-Strength Cast-Iron Electrode, it is claimed to have high as-welded strength.

●

All-Purpose Welder

The Lincoln Electric Co., Cleveland 17, Ohio, has introduced a new universal, combination arc welder which provides a choice of either a-c or d-c welding current. The new machine, called the "Idealarc," is said to provide an ideal type of welding arc for every type of manual welding application, permitting selection of either ac or dc and either a soft or forceful arc.

The Idealarc machine is available in several combinations. It can be obtained as an a-c welder without the d-c current. To this unit a d-c package can be easily attached in about one hour whenever desired. It is also available as a combination a-c, d-c machine with selection of either current made through a simple twist of a switch handle. Different output capacities of dc and ac can be combined to fit the machine to the job requirements.

For a-c welding jobs the machine provides a single-phase transformer type welder, the circuit of which has, according to the manufacturer, been designed to give greater arc stability than

Automatic welding
at Solar Aircraft

CRUCIBLE stainless steel welding wire

for faster, cleaner, uniform weld quality

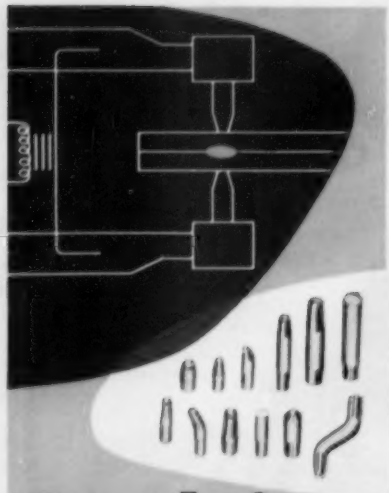
Save time and money . . . get better weldments with Crucible stainless steel welding wire. For Crucible welding wire is made by *steelmen* for optimum performance on the job.

You can get these superior welding wires for any type of automatic welding . . . in most stainless grades and sizes . . . and in spools, coils and cut lengths.

They're all quickly available through your nearby Crucible warehouse. Next time you need stainless welding wire call Crucible. *Crucible Steel Company of America, Henry W. Oliver Building, Pittsburgh 30, Pa.*

CRUCIBLE first name in special purpose steels

Crucible Steel Company of America



Good tips
for resistance-welding machine users

AMPCO WELD*
spot-welding tips
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is normally associated with welding transformers. Both voltage and amperage are controlled to give the desired control of arc characteristics. The Idealarc provides continuous current control through a rotating core in a separate control circuit and also has a switch to permit control of the open-circuit voltage. In addition to dual-control of the arc, the Idealarc machine also has an Arc Booster switch for selecting normal or "hot" starting.

For d-c welding, the Idealarc provides d-c welding current through heavy-duty rectifiers. Dual-control of the arc is provided and, in addition, ability to select either normal or "hot" starting with the Arc Booster.

Current models available are 300-, 400- and 500-amp ac combined with dc capacities in 200, 300, 375 and 450 amp.

•

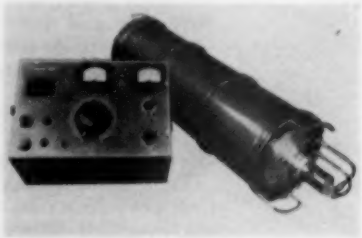
All-Purpose Electrode

The story on the new "Sureweld 'CE' " electrode submitted by the National Cylinder Gas Co. of Chicago and published on page 1206 of the December 1954 issue of THE WELDING JOURNAL contained an error. The first sentence of the third paragraph read: " 'CE' is classed as an AWS 6010 electrode but exceeds specifications for this class in physical properties." Actually, the "CE" electrode is an AWS class E-6013 electrode—not E-6010 as stated in the original release.

•

X-Ray Unit

A new portable unit for complete circumferential X-ray of weldments and construction of pipe lines, castings, boilers, tanks and similar equipment in the field and in industrial plants and on ships has been announced by Mitchell Radiation Products Corp., 128 E. Washington St., Norristown, Pa. The



high-powered, lightweight unit, called the X-Tron 180, reportedly contains a specially designed transformer and X-ray tube and a high-voltage generator that delivers 180 kv and 10 ma continuously. The unit will penetrate 2⅛ in. of steel and will traverse a pipe line greater than 12 in. in diameter.

Hard-Facing Electrode

The new Mir-O-Col No. 2 hard-facing electrode is recommended by its manufacturer for rebuilding heavy-duty equipment where severe impact and abrasion in combination are paramount factors. Free samples of this electrode, along with descriptive literature, are available on request to Mir-O-Col Alloy Co., Inc., 312 North Avenue 21, Los Angeles 31, Calif.

•

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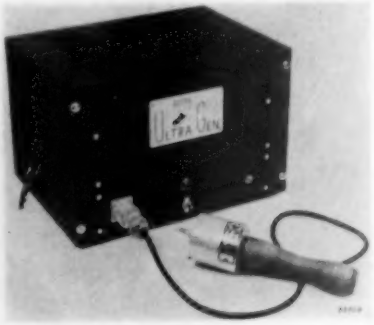

For complete information, write to Alcar Instruments, Inc., Division of Eastern Electro-Sonic Industries Corp., 595 Madison Ave., New York 22, N. Y.

•

Electrodes for Underwater Operations

Two new electrodes for underwater work, the Pacific UC electrode for underwater cutting and Pacific UW for underwater welding, have been placed on the market by Pacific Welding Alloys of Los Angeles, Calif.

Pacific's UC electrode is a steel tube 14 in. long with a $^{5}/_{16}$-in. OD and 0.112 in. ID, with an extruded flux coating which makes it completely waterproof. Pacific's UW electrode is all steel with an extruded and waterproofed coating developed especially for underwater welding.

Free samples of both electrodes, along with descriptive literature, are available by writing Pacific Welding Alloys Mfg. Co., 310 North Avenue 21, Los Angeles 31, Calif.

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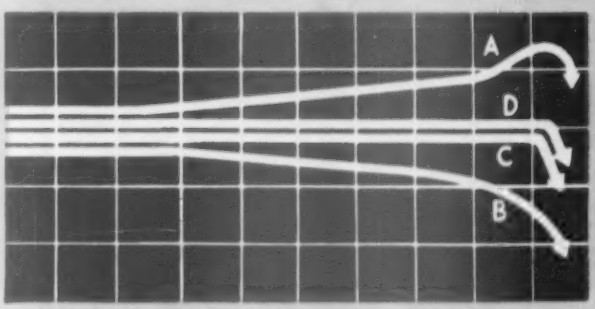

Performance Graph of Various Types of Regulators

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REMINDER TO AUTHORS FOR 1955 AWS NATIONAL FALL MEETING

Authors who plan to propose papers for the 1955 AWS National Fall Meeting in Philadelphia are reminded to forward their abstracts to National Headquarters before the March 15th deadline to assure full consideration. If needed, additional copies of the Abstract Form may be obtained by writing to National Secretary Joseph G. Magrath, American Welding Society, 33 West 39th Street, New York 18, New York.

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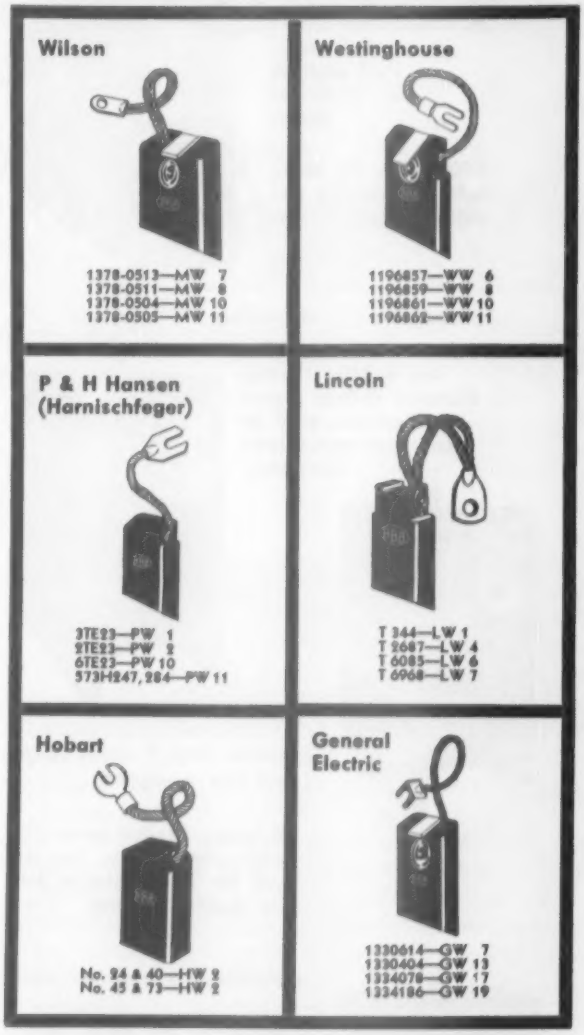

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WELDING RESEARCH COUNCIL
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SUPPLEMENT TO THE WELDING JOURNAL, MARCH, 1955

THE EFFECT OF MICROSTRUCTURE ON THE MORPHOLOGY OF FRACTURE—PART I

An investigation on the effect of pearlitic microstructures in the morphology of fracture and a theoretical explanation on the deformation behavior of pearlite

BY J. C. DANKO AND R. D. STOUT

Introduction

Pearlite, a lamellar aggregate of ferrite and cementite that results from the eutectoid reaction has been the subject of many investigators. Foremost among these investigators were Mehl and associates.[1–3] Their fundamental research on the lattice relationships in the decomposition of austenite to pearlite, the mechanism of pearlite nucleation and growth and the interlamellar spacing of pearlite revealed for the first time a distinct picture of the pearlite reaction.

Attention was also focused on the mechanical properties of pearlite microstructures produced in plain carbon steel of eutectoid composition. Gensamer, *et al.*[4, 5] investigated the mechanical properties of pearlite produced by the isothermal decomposition of austenite over a wide range of decomposition temperatures. Tensile and yield strength increased linearly with decreasing reaction temperatures but the ductility, i.e., the reduction of area and elongation showed a maximum at approximately the middle of the reaction temperature over which the pearlite was formed. More recently, however, investigations have been undertaken to evaluate the notch toughness of pearlitic steels. Rinebolt and Harris[6] determined the effect of alloying elements on the notch toughness of pearlitic steels at a constant pearlite spacing. The influence of pearlite spacing on notch toughness was investigated by Gross and Stout.[7] Their results revealed that in fine-grained steel the best notch toughness occurs at the coarsest pearlite spacing whereas coarse-grained pearlites exhibited a maximum in notch toughness. On the contrary, Rinebolt[8] found that the transition temperature based on a 20 ft-lb criterion decreased with increasing hardness, implying finer pearlite spacings.

However, more recent experiments by Gross and Stout[9] confirmed their earlier results. This investigation on the effect of pearlitic microstructures on the morphology of fracture conducted simultaneously also corroborates their results. In addition, a theoretical explanation on the deformation behavior of pearlite is presented.

Experimental Techniques

The chemical analysis of the plain carbon steel of eutectoid composition is presented in Table 1. Speci-

Table 1—Chemical Analysis of Steel

C, %	0.80
Mn, %	0.74
P, %	0.021
S, %	0.04
Si, %	0.17

J. C. Danko and R. D. Stout are associated with Lehigh University, Department of Metallurgy, Bethlehem, Pa.

Presented at the Thirty-Fifth National Fall Meeting, AWS, held in Chicago, November 1-5

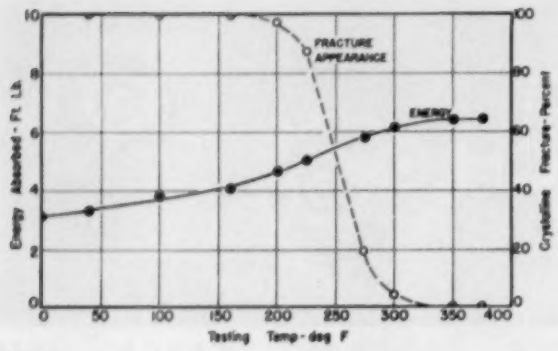

Figure 1

men blanks $^1/_4$ x $^1/_2$ x 3 in. were cut from $^1/_2$- x 3-in. rolled stock. The specimen blanks were austenitized at 2000° F (1093° C) and isothermally transformed at 1290° F (698° C). This heat treatment produced a coarse prior austenitic grain size of ASTM No. 0 and a coarse pearlitic microstructure. This coarse-grained microstructure facilitated micrographic and fractographic observations.

Subsized Charpy V-notch bars with a breadth of 0.197 in. were machined from the heat-treated specimen blanks. The subsized bars were tested in a Baldwin Southwark impact tester of 16 ft-lb capacity and a striking velocity of 11.3 fps. After the transition temperature curve was obtained, pre-polished and etched Charpy bars were tested above, in and below the transition temperature range. Prepolished and etched bars were employed to detect the mode of deformation and the presence of microcracks. If unpolished bars were used and subsequently ground and polished all traces of deformation by slip and certainly the microcracks would be removed. Supplementary information on the deformation of pearlite was procured by testing prepolished and etched flat tensile specimens, and small compression specimens. Specimens with finer pearlite spacing tested in compression were produced in the investigation conducted by Gross and Stout.[9]

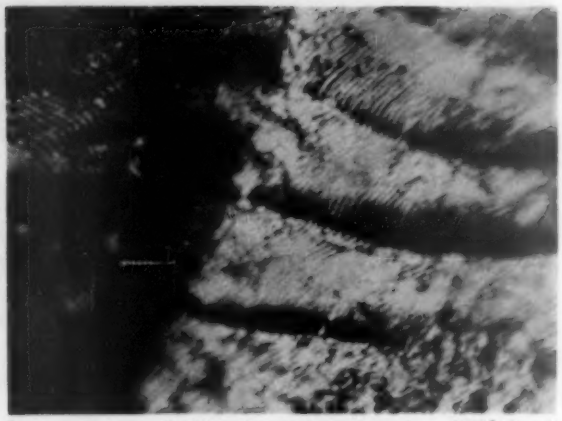

Fig. 2 Extreme kinking in specimen tested at 375° F. X 1500. Polarized light. Picral etch

Results and Discussion

Coarse Pearlite

The transition temperature curve for the coarse pearlite microstructure is illustrated in Fig. 1. Microscopic observations on a prepolished and etched specimen tested above the transition temperature range (375° F) showed pronounced deformation. However, the mode of deformation appeared quite different from ordinary deformation by slip and twinning. The structure shown in Fig. 2 reveals large amounts of buckling or kinking. From this photomicrograph, it appears that the cementite has actually plastically deformed in a kinking-type deformation process. Deformation by kinking, also occurred on the specimen tested in the transition temperature range (225° F). At this temperature the breadth of the kinks decreased and some extremely sharp kinks that evidently developed microcracks were observed. This is illustrated in Fig. 3. Additional evidence of the sharp kinking was obtained on the specimen tested below the transition temperature range (−100° F). Although kinking occurred, the breadth of the kinks were less than those observed in the previous specimens. The sharp kinks are shown in Fig. 4. Since the appearance

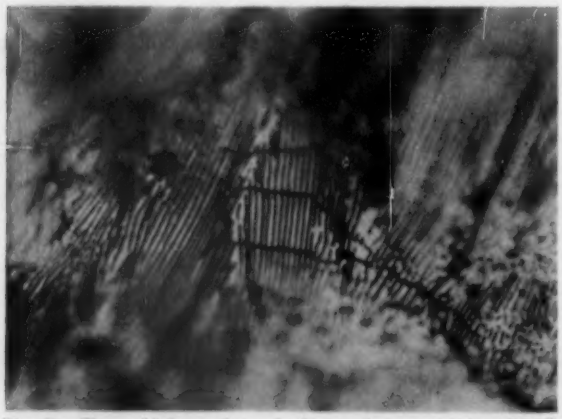

Fig. 3 Sharp kinks and cracks in specimen tested at 225° F. X 1500. Picral etch

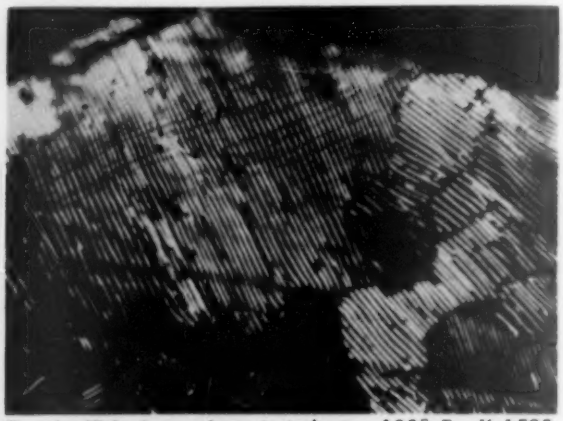

Fig. 4 Kinks in specimen tested at −100° F. X 1500. Polarized light. Picral etch

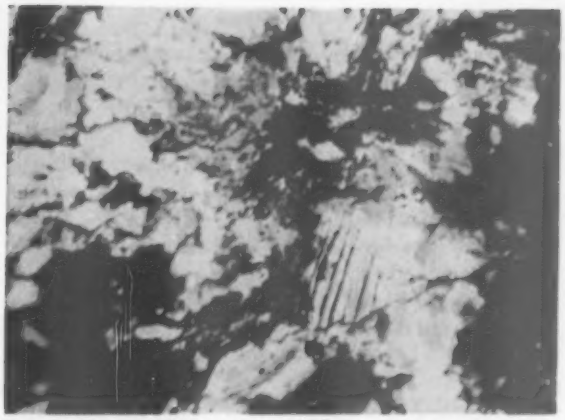

Fig. 5 Pearlite formed at 1100° F. Kinks produced by compression. X 1500. Polarized light. Picral etch

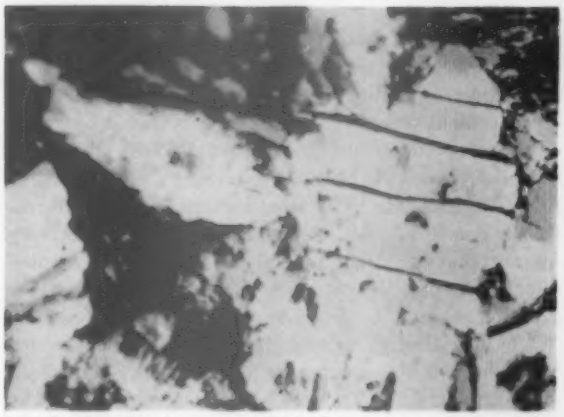

Fig. 6 Pearlite formed at 1185° F. Kinks produced by compression. X 1500. Picral etch

of the sharp kinks is associated with the onset of brittle failure, they may be responsible for initiating fracture.

It appears that deformation at temperatures at which the kinking is of the gradual type results in a normal ductile behavior. As the temperature of deformation is lowered the breadth of the kink bands is appreciably reduced. With the manifestation of extremely sharp kinking, brittle failure is encountered. The abruption of the cementite by the sharp kinks initiates fracture. This fact was confirmed by extensive microscopic observations and was illustrated by the photomicrograph of Fig. 3. Thus, in the coarse pearlite the mode of deformation is essentially one of plastic and elastic kinking and brittle failure is definitely associated with the presence of sharp kinks. This same behavior should also be manifested in pearlites of finer interlamellar spacing.

Fine Pearlites

Deformation of the finer pearlites was produced by simple compression of prepolished and etched microspecimens. Pearlite formed by the isothermal decomposition at 1100° F (593°C) deformed by the kinking mechanism observed in the coarse pearlite. However, the kinks were extremely sharp and much smaller in length as shown in Fig. 5. With increasing lamellar spacing of the pearlite, the kink bands become greater in breadth and length. This is illustrated in Fig. 6, for the compressed specimen of pearlite formed at 1185° F (640° C). Pearlite formed at 1255° F (679° C) reveals greater extent of kinking than the previous finer pearlites and Fig. 7 shows this pronounced kinking. The broadest kink bands were observed in the coarsest pearlite that was formed at 1300° F (704° C). Thus, it is evident that the deformation of pearlite by a kinking mechanism is influenced by testing temperature and the interlamellar spacing of the pearlite.

Inasmuch as the finer pearlites produce much sharper kinks than the coarser pearlites, the coarser pearlites should exhibit better notch toughness. In order to ascertain this view, the extent of microcracking was approximated for the various pearlites compressed at approximately 5% strain. Although the data are very qualitative, the formation of microcracks increased as the pearlite spacing decreased. This, of course, was anticipated since the sharper kinks of the fine pearlite result in cracks more readily than the larger gradual kinks associated with the coarse pearlite. That pearlites of coarse interlamellar spacing exhibit superior notch toughness was reported by Gross and Stout.[7, 9]

Deformation of Pearlite

Since pearlite is an aggregate of cementite and ferrite, the deformation of this structure is somewhat complex. The cementite which is coherent with the ferrite restrains the ferrite from deforming by a slip process. Consequently, the ferrite remains elastic until the cementite deforms elastically or plastically. The action of the elastic ferrite may be to furnish an environment in which plastic deformation of the ordinarily brittle cementite becomes possible. Bridgman[10] has shown that ordinarily brittle materials behave in a ductile manner in compression and tension test conducted under high hydrostatic pressures. If the ferrite of the pearlite can provide conditions analogous to a hydrostatic pressure, the cementite may then be expected to undergo greater elastic strains and may possibly behave

Fig. 7 Pearlite formed at 1255° F. Kinks produced by compression. X 1500. Picral etch

in a ductile fashion. On this basis the deformation of the pearlite and the plastic behavior of the cementite could be explained.

Kinking of pearlite involves a cooperative movement of the cementite and ferrite. This movement generally occurs over the entire pearlite colony and may often extend across other colonies providing the cementite is properly oriented. The most favorable kinking direction is one in which the cementite is oriented approximately perpendicular to the kink. However, small deviations from the perpendicular have been observed. Kinking occurs in tension as well as compression. While it is difficult to visualize a kinking phenomenon in tension, kinking has been reported in tension in single crystals of zinc.[11] The tensile kinks were attributed to nonuniform distribution of flow. A similar mechanism may be operating in the polycrystalline aggregate of pearlite in which complex stresses exist. However, more observations are necessary to corroborate this viewpoint.

Kinking of pearlite in compression and tension resulted in the formation of microcracks. These cracks were initiated in the kinked areas and propagated along the kink direction. The propagation is consistent with the observations on the fractured pearlite, i.e., cleavage occurs with the cementite approximately perpendicular to the cleaved surface. Moreover, the cleavage facets appear to originate from individual pearlite colonies.

Since cleavage appears to occur in single pearlite colonies, the pearlite colony size may exert an influence on the notch toughness Kinking is generally confined to individual pearlite colonies; therefore, less deformation results in the pearlites of small pearlite colonies. In addition, the pearlitic structures with fine pearlite colonies are those formed at lower reaction temperatures where finer pearlites are produced. Consequently, smaller and sharper kinks results in the fine pearlites and reduce the amount of deformation. Therefore, the mode of deformation and the pearlite colony size should be considered in the behavior of notch toughness of pearlitic structures.

Conclusions

1. Pearlite deforms by a kinking type process.

2. Deformation above the transition temperature occurs by the formation of very broad kinks. In the transition temperature and below this range the breadth of the kinks decreases and extremely sharp kinks appear. These sharp kinks initiate fracture and are directly associated with the brittle behavior.

3. Finer pearlites deform by a kinking mechanism in which the kinks are much sharper. The breadth of the kinks decreases with decreasing interlamellar spacing of the pearlite. Consequently, the finer pearlites may be expected to exhibit a lower notch toughness.

4. The length of the kink band is limited to the size of the pearlite colony. However, the kinks will extend into adjoining colonies when the cementite lamellae are of proper orientation, i.e., perpendicular to the kink direction.

Acknowledgment

This investigation was made possible by a grant-in-aid from the University Research Committee of the Welding Research Council, for which the authors are deeply grateful.

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BULLETIN 21—COATING MOISTURE INVESTIGATIONS

The Welding Research Council has issued Bulletin 21—Coating Moisture Investigations of Austenitic Electrodes of the Modified 18-8 Type, by K. P. Johannes, D. C. Smith and W. G. Rinehart. This report gives rates of moisture absorption and desorption of austenitic welding electrode coatings of various temperatures and relative humidities. Copies may be purchased at $1.00 each through the AMERICAN WELDING SOCIETY, 33 West 39th Street, New York 18, N.Y.

THE EFFECT OF MICROSTRUCTURE ON NOTCH TOUGHNESS PART II

Information on the variables affecting the notch toughness of unhardened plain carbon steels

BY JOHN H. GROSS AND ROBERT D. STOUT

The need for this detailed investigation springs from the importance of notch toughness in a wide variety of unhardened or partially hardened plain carbon and alloy steels ranging from rail steel to low-carbon structural steels. In the case of rail steel, for example, the improved notch toughness of fine pearlite produced by oil quenching and tempering as opposed to a softer, coarser, pearlite is not fully understood. Nor can the effects on notch toughness be predicted for the complex structural changes accompanying the welding of low-carbon steels. The importance of these questions is attested by the extensive investigations on the structural steel used in engineering structures which have been found to be susceptible to brittle failure.

In an investigation directed at evaluating the effect of high temperature transformation products on notch toughness, Grossman[1] concluded that an increase in the fineness of pearlite of a hypoeutectoid steel resulted in an improvement in notch toughness. Unfortunately neither the quantity nor type of proeutectoid ferrite was held constant as the pearlite spacing was varied. Thus, while this conclusion may be valid for the aggregate behavior of ferrite and pearlite, the part contributed by changes in the character of the ferrite is not known. Obviously, the manifold variations of microstructure possible in steel present a complex problem. One must consider such variables as pearlite spacing, pearlite carbon content, pearlite colony size, spheroidization, ferrite grain size, ferrite type, presence of other transformation products including bainite and martensite, austenitic grain size, as well as embrittling phenomena. These can be studied only by systematic variation and control. This investigation has been undertaken with the hope that the independent contribution of each variable of microstructure can be evaluated and ultimately that the notch toughness of combined aggregates can be explained or even predicted from microstructural observation.

John H. Gross is Instructor and **Robert D. Stout** is Professor of Metallurgy at Lehigh University.

Presented at the Thirty-Fifth National Fall Meeting, AWS, held in Chicago, November 1–5.

Table 1—Chemical Composition, %, of Steels

	C	*Mn*	*P*	*S*	*Si*
C1080	0.84	0.74	0.014	0.034	0.22
C1090	0.95	0.83	0.020	0.037	0.22

As a logical starting point, Part I of this investigation[2] was limited to a study of eutectoid plain carbon steels. Unfortunately the results of this investigation could not be reconciled with those obtained later by Rinebolt.[3] Since a thorough understanding of the simplified eutectoid problem is essential to the extension of the problem to hypoeutectoid steels, additional information on this subject is presented here. While the previous results have for the most part been substantiated, improved experimental techniques have revealed the existence of additional variables which exercise an influence on the behavior of pearlite. Moreover, a corollary investigation[4] conducted separately at this laboratory on the micromechanism of deformation and fracture has provided support to these conclusions. However, the work reported here opens further areas which require investigation.

Experimental Procedure and Results

Steels

The steels investigated were a plain carbon eutectoid and a slightly hypereutectoid steel of chemical composition shown in Table 1. The hypereutectoid steel was included because of the existence of traces of proeutectoid ferrite in the eutectoid steels of the current and previous investigation.

Specimen Preparation

The standard V-notch Charpy bar was used for transition temperature determination. The steel stock was obtained in the form of $^{7}/_{16}$-in. bar stock. Lengths of 3 in. for Charpy specimens and of 4 in. for tensile specimens were cut from the bar stock and drilled at one end to permit suspension in the liquid heating media. Upon completion of heat treatment 0.022 in. was ground from each side of the Charpy specimen which was then cut to standard length and notched. Half-size tensile

specimens were machined from the heat-treated 4-in. bars.

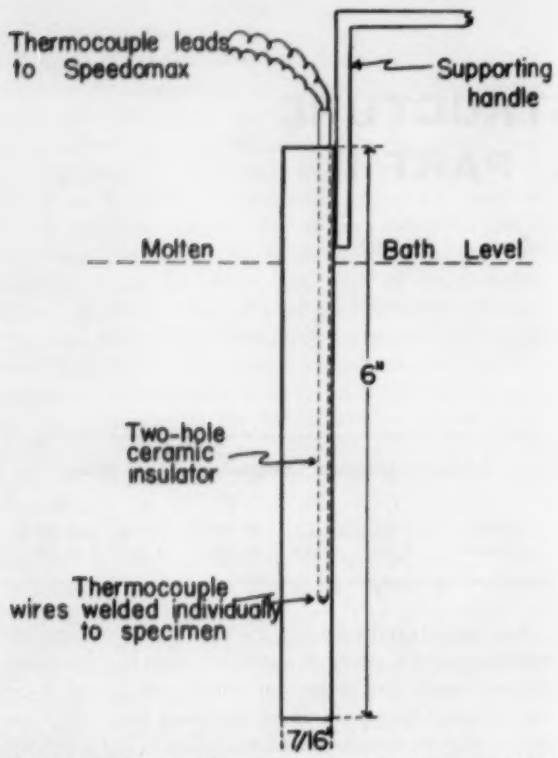

Fig. 1 Specimen and equipment used in determining the mean transformation temperatures

Heat Treatment

For each condition a series of 14 Charpy bars and two tensile specimens were treated simultaneously. Specimens were austenitized at 1450° F (787° C) or 1900° F (1036° C) to produce fine- or coarse-grained austenite followed by the appropriate subcritical isothermal transformation to produce the desired type of pearlite. All heat treatment was performed in molten lead except the 1900° F austenitizing which was done in molten salt. All specimens were normalized at 1600° F (871° C) prior to austenitizing and subcritical isothermal transformation. All specimens were brine quenched upon completion of isothermal transformation. To avoid the possible formation of small quantities of martensite, specimens transformed just below the lower critical for long periods of time were permitted to cool to 1250° F (675° C) prior to brine quenching. Due to the size of the specimen, instantaneous cooling to the desired subcritical transformation temperature could not be obtained. Therefore, transformation, except at temperatures approaching the lower critical, occurred at a temperature range substantially higher than the lead bath temperature. Moreover, heating of the specimen due to recalescence resulted in an increase in temperature during transformation particularly at the lower temperatures and shorter times. In order to determine accurately the actual mean reaction temperature, measurements of this temperature were made by employing the experimental setup shown in Fig. 1 to record the progress of transformation by isothermal treatment

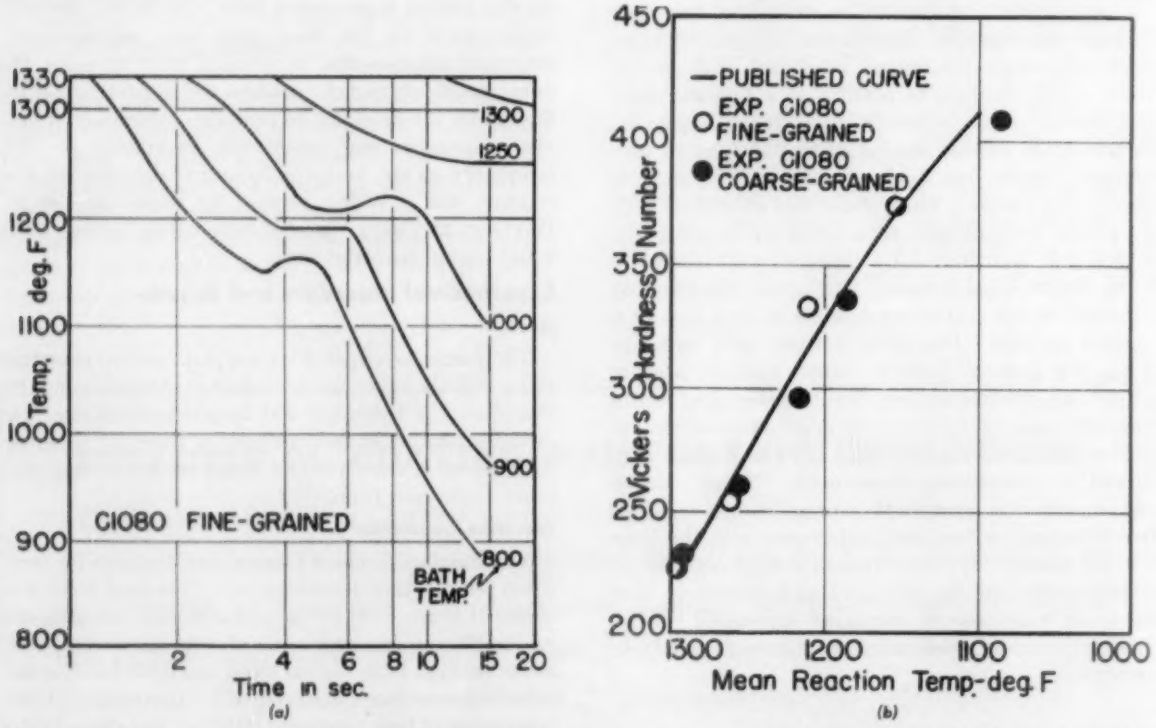

Fig. 2 Cooling curves showing the mean transformation temperatures and resulting hardnesses for C1080 steel quenched to indicated bath temperatures

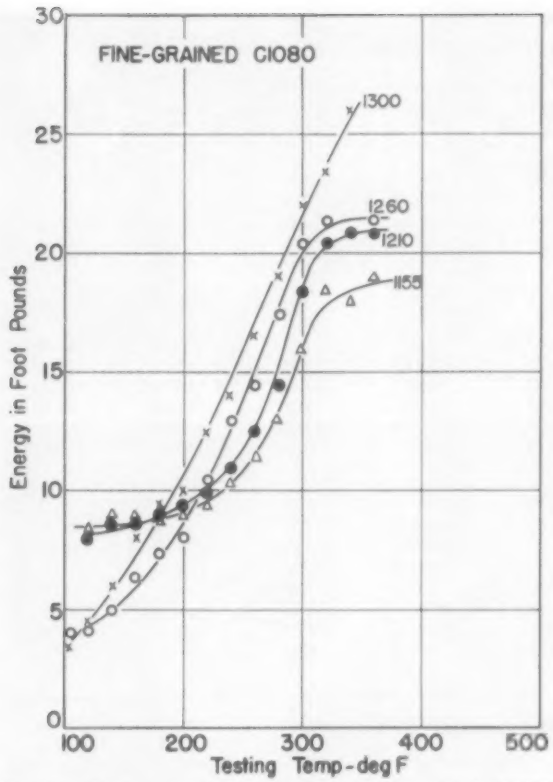

Fig. 3 Charpy transition curves for fine-grained C1080 steel transformed at the indicated temperatures

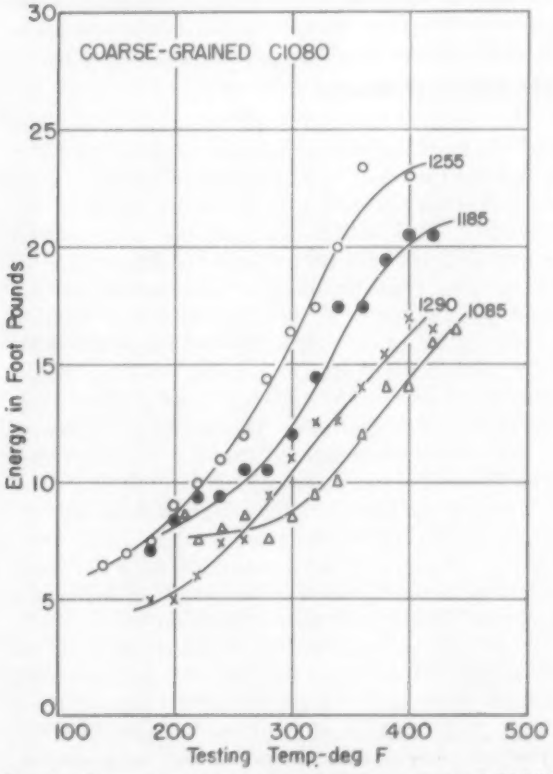

Fig. 4 Charpy transition curves for coarse-grained C1080 steel transformed at the indicated temperatures

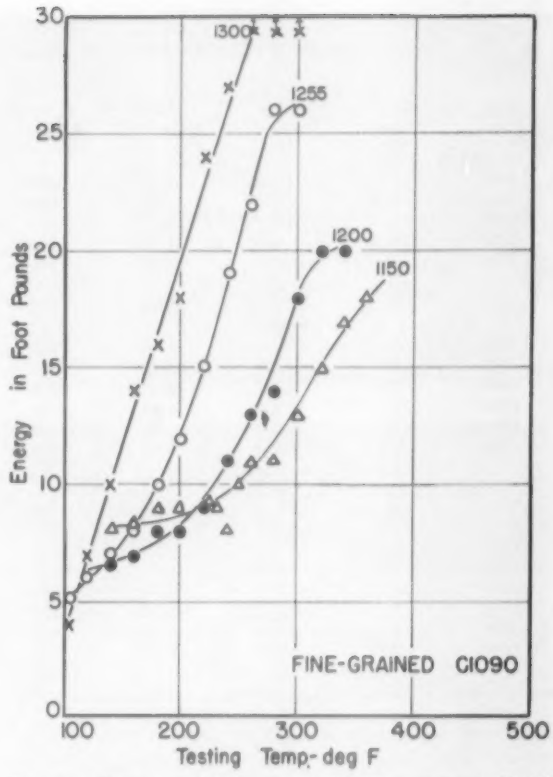

Fig. 5 Charpy transition curves for fine-grained C1090 steel transformed at the indicated temperatures

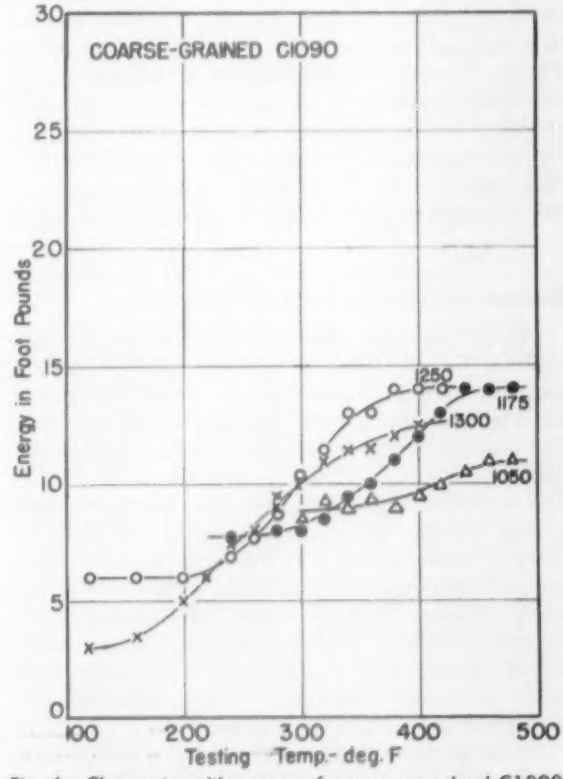

Fig. 6 Charpy transition curves for coarse-grained C1090 steel transformed at the indicated temperatures

Table 2—Test Results for C1080 and C1090 Fine- and Coarse-Grained Steels

Aust. temp.	*Grain size*	*Bath temp., ° F*	*Transf. temp., °F*	*Vickers hardness*	*10 ft-lb trans. temp., °F*	*50% fibrous fracture trans. °F*	*Ultimate tensile strength, kips*	*Elong., %*	*Red., %*
				C1080					
1450	6½	1300	1300	222	195	280	100	21.4	28.6
1450	6½	1250	1260	254	215	295	126	15.9	34.2
1450	6½	1000	1210	334	220	310	166	15.4	43.8
1450	6½	800	1155	374	235	320	183	13.0	42.0
1450	6½	800 & 1300	1155	294	145	260	142	19.8	50.7
1450	6½	1300 & 800	1300	218	175	270	104	22.0	39.6
1900	3	1290	1290	232	295	345	111	9.4	10.6
1900	3	1250	1255	260	225	280	125	13.1	21.2
1900	3	1200	1215	294	250	300	143	10.8	21.5
1900	3	1100	1185	337	260	315	162	11.6	25.4
1900	3	800	1085	408	330	380	187	8.7	18.8
1900	3	750	1000	410	310	400	199	10.4	27.5
				C1095					
1450	7½	1300	1300	218	140	220	105	25.4	52.2
1450	7½	1250	1255	265	180	250	126	21.0	46.6
1450	7½	1000	1200	348	230	285	166	15.6	40.4
1450	7½	750	1150	410	250	330	196	12.4	36.6
1900	4	1300	1300	244	300	340	116	14.0	18.8
1900	4	1250	1250	294	300	330	138	14.4	17.8
1900	4	1100	1175	360	360	410	178	10.2	17.8
1900	4	700	1050	458	415	465	218	7.6	16.1

at subcritical temperatures for the two steels and for the two grain sizes employed. Typical cooling curves for the C1080 steel are shown in Fig. 2*a*. The mean reaction temperature was taken as the average of the minimum value prior to recalescence and the maximum value due to recalescence. The validity of this method in determining mean reaction temperatures is shown in Fig. 2*b*. It will be noted that the actual hardness of the Charpy specimens agrees reasonably well with published isothermal transformation hardnesses[5] for a similar steel at corresponding reaction temperatures. The complete list of lead bath temperatures and resulting mean reaction temperatures is given in Table 2.

Specimen Testing

Charpy specimens were heated to the desired temperature by immersion in glycerin and tested in a standard impact testing machine. Both the energy absorbed and the fracture appearance were recorded. The ductility transition behavior for fine- and coarse-grained C1080 and C1090 is depicted in Figs. 3–6. The 10 ft-lb transition temperature as well as the 50% fibrous fracture transition temperature are listed in Table 2. Also reported in this table is the average hardness at a point corresponding to the root of the notch. Vickers hardness values were determined on the cross section of all Charpy specimens at the center and on either side at notch depth. The results indicated satisfactory uniformity within each bar and among the various bars of any series.

Tensile properties obtained by testing standard half-size tensile bars in duplicate for each heat treatment are also included in Table 2.

Metallographic Examination

Austenitic grain size determinations were made on a number of hardened and fractured specimens of each steel by comparison with fracture standards and are recorded in Table 2. Photomicrographs of all structures resulting from the various treatments were made. Figure 7 includes some typical examples.

Discussion of Results

The variation in the 10 ft-lb transition temperature with the mean reaction temperature is plotted in Fig. 8. It is noted that the notch toughness of the fine-grained steels decreases linearly with decreasing reaction temperature. Thus, it appears that the soft, coarsely spaced pearlites have better notch toughness than the harder, finer types produced at lower subcritical temperatures. However, these steels which were obtained from standard production heats were found to exhibit under certain circumstances a strong tendency toward spheroidization even for the short time of transformation at lower temperatures. This tendency is apparent in the photomicrograph of the fine-grained C1090 (Fig. 7*b*) steel transformed at 1300° F (704° C). For this reason the authors have refrained from any attempts to correlate transition temperature with pearlite spacing. However, if notch toughness is to be related to microstructure a means of quantitatively expressing structural variation will have to be found.

It will also be noted in Fig. 8 that the slope of the fine-grained C1080 differs from the fine grained C1090 and also the straight line portions of the coarse-grained steels. This results in reduced notch toughness of the C1080 at least at the higher reaction temperatures. It is believed that this reduced notch toughness can be accounted for by the lesser spheroidization of the fine-

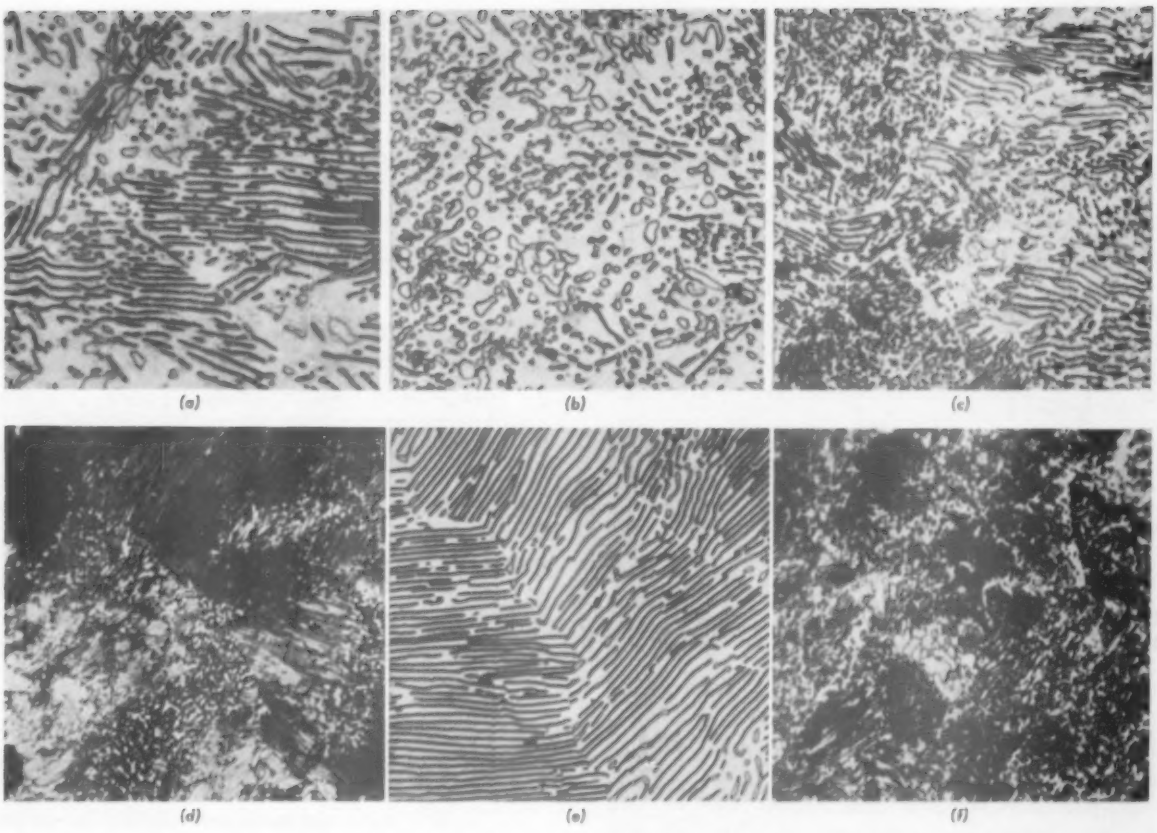

Fig. 7 Typical microstructures showing the distribution of ferrite and carbide from various subcritical transformations

grained C1080 at these temperatures. Comparison of Figs. 7*a* and 7*b* demonstrates the microstructural differences.

A very similar behavior is also observed in the coarse-grained notch toughness behavior. Here it is noted that, as in fine-grained steels, the basic behavior is an increase in transition temperature with increasing reaction temperature and at about the same rate in both steels. The apparently anomalous behavior at the highest reaction temperature may also result from spheroidization. Figures 7*c* and 7*d* are typical of the structures represented by the behavior at the lower reaction temperatures. Spheroidization in this photomicrograph is quite noticeable. In contrast Fig. 7*e* exhibits almost perfect lamellar structure with no detectable spheroidization. Thus, the decreased notch toughness of the coarse-grained specimens transformed at high temperatures may be associated with freedom from spheroidization.

Embrittlement upon quenching from the various subcritical temperatures was suggested as a possible source of variations in notch toughness. To examine this possibility two heat treatments of the C1080 steel were performed: (1) transformation at 1300° F (704° C) plus tempering at 800° F (426° C) for 10 min and (2) transformation at 1155° F (618° C) which required a lead bath temperature of 800° F plus tempering at 1300° F for 10 min. Each operation was followed by the usual water quench. The 800° F tempering of the specimens reacted at 1300° F showed negligible improvement in notch toughness. On the other hand the 1300° F tempering of the specimens reacted at 1155° F resulted in a decrease in transition temperature from 235° F in the untempered specimens to 145° F in the tempered ones. The tempering for 10 min caused a decrease in hardness from 374 VPN to 294 VPN and very marked spheroidization as shown in Fig. 7*f*. Since a marked change in hardness and structure occurred, the changes cannot be ascribed to embrittlement, particularly since the specimens reacted at 1300° F showed a very slight improvement when tempered at 800° F. However, the results emphasize the pronounced improvement that can result due to appreciable spheroidization of nominally lamellar pearlitic structures. Spheroidization may also explain the observed advantage to notch toughness of rail steel when oil quenched and tempered.

Another factor that may contribute to the poor notch toughness of coarse-grained steels at the highest reaction temperature is pearlite colony size. It was noted that the crystalline fractures of the coarse-grained steels were noticeably coarser for the higher reaction temperatures. The fine-grained steels on the other hand showed very little variation in fracture appearance. Since no satisfactory metallographic technique for clearly outlining pearlite colonies could be developed, reliable quantitative measurements could not be made.

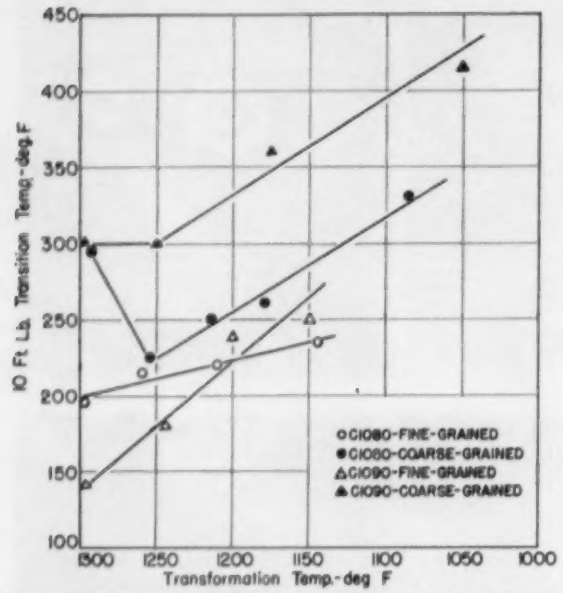

Fig. 8 Relations of Charpy transition temperatures to temperatures of transformation of C1080 and C1090 steels

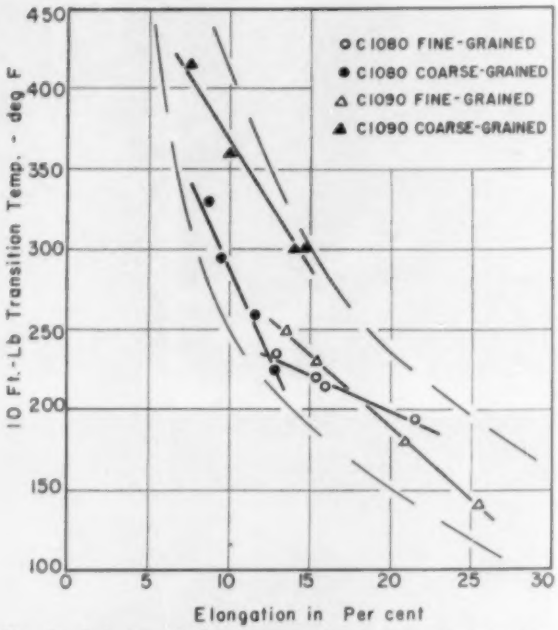

Fig. 9 Relation of Charpy transition temperatures to tensile elongation of C1080 and C1090 steels

However, this variable may be found to exert an appreciable influence when properly evaluated.

As previously reported, a correlation exists between transition temperature and tensile ductility. Figure 9 shows the general improvement in notch toughness with increasing tensile elongation. More interesting however is the division of the correlation into families of linear relationships on the basis of composition and grain size. Thus, the qualitative notch-toughness-ductility relationship appears reasonably quantitative when limited to a given type of steel. Reduction of area yields a similar qualitative relationship but indicates no family relationship as observed with elongation.

Conclusions

1. Generally the notch toughness of eutectoid steel decreased as the transformation temperature was lowered and the fineness of the microstructure increased.

2. For the coarse-grained series, a minimum in the transition temperature was observed with transformation at 1250° F. The tendency for toughness to decrease when transformation occurred at 1300 ° F appeared to be associated with less spheroidization at this temperature.

3. In the commercial steels tested in this investigation a marked tendency toward spheroidization was observed. Spheroidization favors notch toughness. Thus, while reduced pearlite spacing obtained at lower reaction temperatures lowers notch toughness, any concomitant spheroidization will modify this behavior.

4. Tensile ductility also decreases with lower transformation temperatures. For a given steel and grain size, tensile elongation showed in these tests a linear relation with Charpy transition temperature, but reduction of area failed to show this type of relation.

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STEEL

The WELDING RESEARCH COUNCIL has just issued Bulletin No. 20 consisting of three timely papers by Charles M. Parker, Assistant Vice-President, American Iron and Steel Institute. The titles of these papers are:

1. High-Strength, Low-Alloy Steels
2. Analytical Chemistry Fosters Progress in Steelmaking
3. Steel Quality

All three papers are of interest to the welding industry and are written from a practical viewpoint. Copies of Bulletin No. 20 may be purchased through the American Welding Society, 33 West 39th Street, New York 18, New York, at $1.00 per copy.

SEAM WELDING LOW-CARBON STEEL

Development of basic procedure data for the fabrication of pressure tight lap seam-welded joints in low-carbon steel sheets

BY M. L. BEGEMAN AND GENE C. WALKER

The resistance seam welding of low-carbon sheet steel has been successfully accomplished for some time and finds use in many industrial applications. Such welds are normally used in two basic ways: they are stressed in shear where a continuous joint is desired, and, they are made in assemblies which should be pressure tight. This study will be concerned only with the fabrication of pressure-tight lap joints made with low-carbon sheet steel. Although low-carbon sheet steel has a wide welding range it is important that proper machine settings be known for a given thickness material to obtain good welds, free from porosity, pressure tight and of adequate strength.

Material and Its Preparation

This work covers the resistance seam welding of low-carbon, pickled and oiled, hot-rolled, rimmed steel, commonly designated as SAE 1010.* Three thicknesses were used of the following analyses:

Gage, in.	%			
	C	*Mn*	*P*	*S*
0.031	0.08	0.39	0.008	0.029
0.049	0.07	0.41	0.012	0.037
0.078	0.08	0.39	0.010	0.034

Some difficulty was experienced in obtaining consistent welds in the as-received condition due to variable surface conditions. This problem was solved by lightly wire brushing the test specimen and then wiping with naptha to remove all grease and dirt. Degreasing with acetone also gave consistent values of contact resistance. Average contact resistances after degreasing for the three thicknesses are:

Gage, in.	*Contact resistance, microhms*
0.031	795
0.049	433
0.078	30

* The sheet steel used in this work was supplied by the United States Steel Co.

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Presented at the Thirty-Fifth National Fall Meeting, AWS, held in Chicago, Nov. 1–5, 1954. This research work was carried out under the auspices of the Resistance Welding Committee of the Welding Research Council under the direction of a Subcommittee consisting of J. J. Riley, Chairman; R. C. Becker, M. L. Begeman, J. Deffenbaugh, R. T. Gillette, I. W. Johnson, D. L. Knight, E. F. Nippes and S. Schneider.

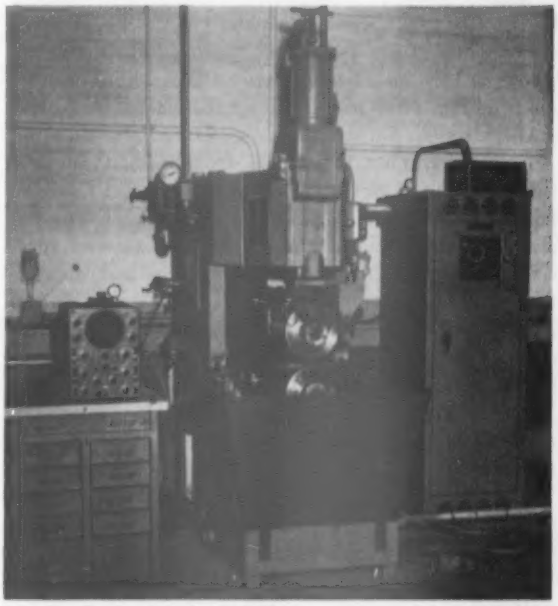

Fig. 1 Taylor-Winfield 250-kva seam welder with G-E control

All specimens for microstructure study, radiographic test and nugget properties were made on strips 1 in. wide. Pillow specimens varied in size from 6 x 6 in. to 10 x 10 in.

Welding Equipment and Instrumentation

The welding in this study was performed on a Taylor-Winfield Type WS CB-22 250-kva friction-driven circular seam welder as shown in Fig. 1.* The electrical control was of the separate mounting type, a General Electric Type CR-7503-B110 control using two size C ignitron power tubes.

The two welding electrodes were each 10 in. in diameter, 1/2 in. thick, RWMA Class 2 Alloy,† having a double bevel machined to provide face width as shown in Fig. 2. For 0.031-in. steel the face width was 1/4 in., for 0.049-in. steel, 5/16 in. and for 0.078-in. steel, 3/8 in. A 1/32 in. radius was provided on each edge.

* The seam welder and electronic control was furnished by RWMA.

† Electrodes were furnished by P. R. Mallory and Co.

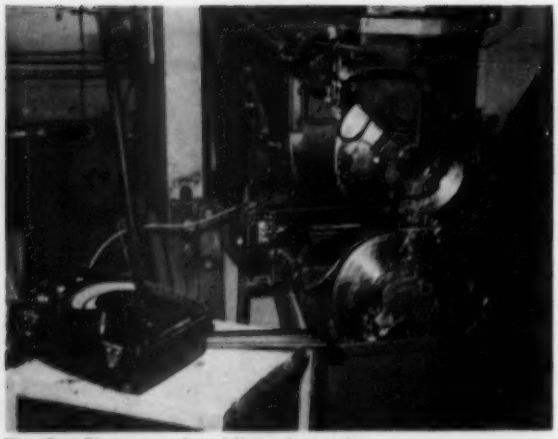

Fig. 2 Close-up of welding electrodes

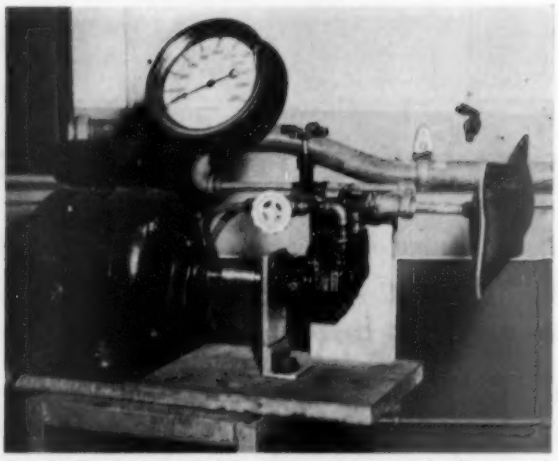

Fig. 3 Equipment used for hydrostatic test of pillow specimens

These tests were all made with a face radius of 6 in. Note in Fig. 2 the guide behind the electrodes to produce consistent pillow specimens.

A series of pillow test specimens were made for each run over a wide range of current values and tested to failure by the equipment shown in Fig. 3. The test is a hydrostatic one in which two sheets of convenient size are seam welded around the edges and then attached to the pumping equipment by means of a small pipe nipple welded to one sheet. If the weld is pressure tight failure will usually occur on an inside edge of the seam.

The instrumentation employed with seam-welding equipment must provide means for measuring the electrode force, welding speed and welding current. To measure the electrode force a General Electric Electrode Force Gage, Model 698780161, was used with suitable adapters to fit it between the wheels. Welding speed was determined by timing electrode rotation with a stop watch.

Perhaps the most difficult quantity to measure is the welding current. The difficulty arises from the use of partial waveforms and interrupted pulses of current. Two methods were employed in current measurement. For work with essentially full wave continuous current and with interrupted current of at least two cycles on time a pointer-stop ammeter and current transformer were employed as seen in Fig. 2. For interrupted current of one cycle on time and for partial wave continuous currents the pointer stop ammeter was used primarily but was checked by use of an air toroid around a current carrying arm of the welder with a suitable integrating circuit and oscillograph as seen in Figs. 1 and 2. The air toroid was calibrated on the machine and provides an accurate means of studying any of the current patterns and waveforms used in seam welding.

Major Welding Variables

The major factors or variables in seam welding are current magnitude, heating time, cooling time, magnitude of force between electrodes and welding speed. The proper control or setting of each variable is necessary for a good weld; however, tests show that for each factor the optimum condition is not a fixed point but rather it exists over a fairly wide range. The correct position in the range is governed by the setting of the other variables. In some cases the setting may be a compromise with the other factors in order to obtain a weld with a certain desired property.

The current magnitude for seam welding is slightly higher than for spot welding due to the shunting effect through adjacent overlapping spots. It is increased as the welding speed is increased or as the thickness of the material to be heated increases. For each given thickness or speed the current values were varied over a wide range in order to determine a weld with satisfactory properties. Heating and cooling times were investigated for each condition and set so as to produce sound welds with satisfactory nugget overlap. The magnitude of the electrode force was controlled so as to produce welds with good penetration and nugget width.

In this series of tests particular emphasis is given to presenting data and schedule for welding speeds higher than those for which data have been available. Limiting factors in going to higher speeds are electrode pickup, possible porosity due to weld cooling without proper electrode force, and welds not having sufficient nugget overlap. The criteria established for satisfactory seam welds are: (1) minimum porosity; (2) sufficient nugget overlap to insure pressure-tight welds; and (3) reasonable allowable deviations in the welding variables to insure uniform results. Schedules are presented for each thickness to aid users of seam-welding equipment in selecting suitable welding conditions for their work.

Procedure

The procedure adopted for determining the most satisfactory welding condition is as follows:

1. A speed is selected which represents a practical working value. For this phase of the work the values in the "Recommended Practices" were used as a guide.

2. Proper cycling is determined by using the recom-

mended force and varying the heat and cool times for a range of current at each cycling pattern. The criteria of proper cycling have been no porosity, and overlap and penetration within the standards established by the committee for this work. Penetration should average from 45 to 50% of original sheet thickness but should be within the limits of 30 to 70%. Overlap should be around 15 to 20%.

3. Using the chosen cycling pattern, welds are made at three to five values of current over a wide range of forces. From the radiographs and data on nugget size, the range of force considered satisfactory is chosen.

4. Using the previously established cycling and force, welds are made over a wide current range to find values which yield suitable nugget overlap, penetration with negligible indentation and no porosity.

5. As each higher welding speed was selected new control values were estimated on the basis of the results from previous runs, but in general the procedure for their final determination was as outlined above.

The type of joint obtained in all cases was a lap joint in which the lap is greater than the wheel face. There was no attempt made to determine the water flow; however, large quantities of water were used above and below the seam in order to sufficiently cool the weld. The control of the water is important as nonuniform cooling results in the production of nonsymmetrical weld nuggets.

Effects of Weld Current

As in all resistance welding the heat produced in the seam welding process is the result of passage of current through the materials being welded. There are three sources of heat. First the contacts between electrodes and the work being welded have resistance. The heat produced at these contacts is kept at a minimum by the use of copper-alloy electrodes and is effectively dissipated by rapid cooling of the electrodes and weld area. In this work external flood cooling with water was used exclusively, as shown in Fig. 2. This is perhaps the most efficient means of cooling as water has a high specific heat and was made to flow rapidly over the electrodes and the hot welded zone. Secondly, the volume resistivity of the metal being welded causes generation of heat. The heat generated near the interface of the pieces being welded contributes to the welding process while that generated near the outside is carried away quickly by the cooling water. Lastly, and of principal importance, is the heating due to the resistance of the contact of the pieces being welded at the interface. This heating plus that due to the internal resistance of the metal can be made to cause melting of the metal in a localized area thus allowing actual fusion of the two pieces to occur at the interface. These heating effects are, of course, proportional to the square of the current and to the resistances involved. When the current is sufficiently large to cause fusion, nuggets of weld metal are formed as seen in the photomacrographs, Figs. 18 through 22. When welding with interrupted current the spacing of these nuggets may be varied from individual welds as in "roll-spot"

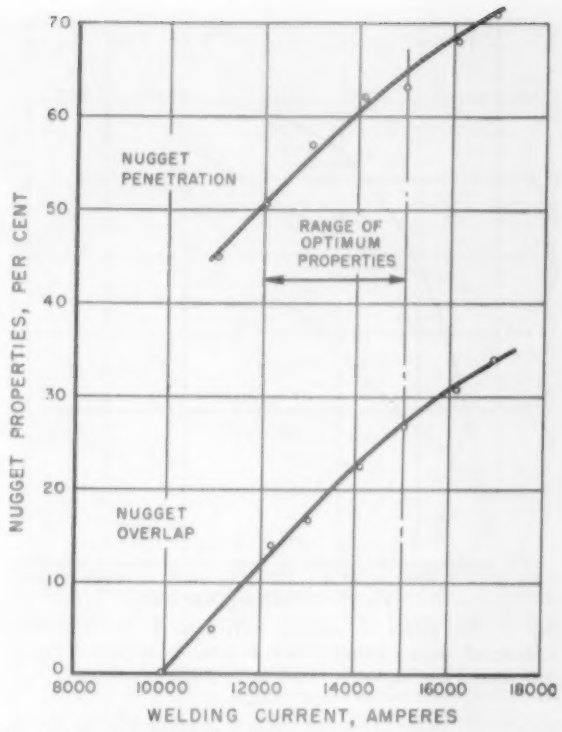

Fig. 4 The influence of welding current on nugget penetration and overlap on welding 0.031-in. steel, heat time 3 cycles, cool time 2 cycles, force 700 lb, and speed 72 ipm

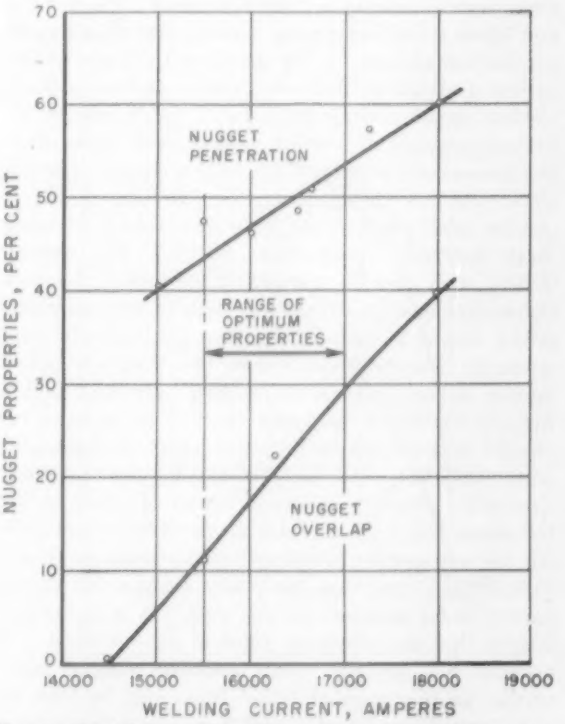

Fig. 5 The influence of welding current on nugget penetration and overlap on welding 0.049-in. steel, heat time 4 cycles, cool time 3 cycles, force 1000 lb, and speed 65 ipm

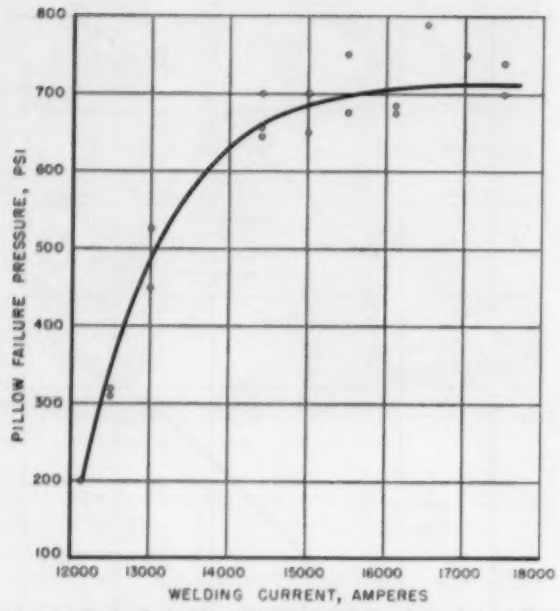

Fig. 6 The effect of welding current on 6- x 6-in. pillow failures of seam welded 0.049-in. steel, heat time 4 cycles, cool time 3 cycles, force 1000 lb, and speed 65 ipm

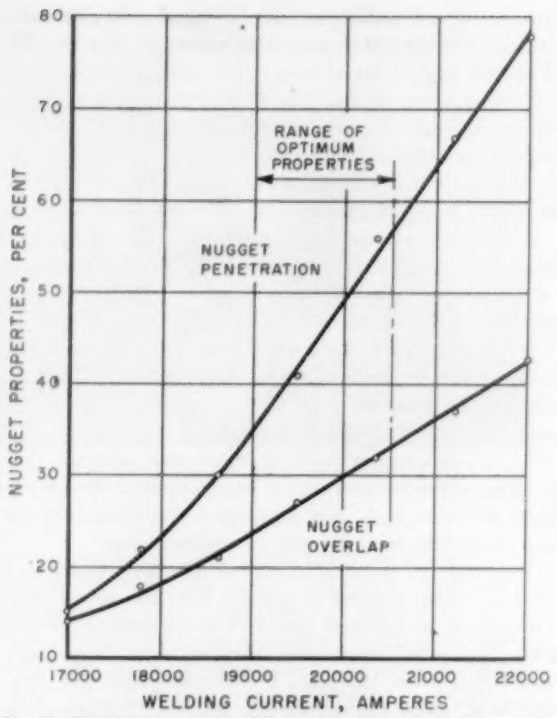

Fig. 7 The influence of welding current on nugget penetration and overlap on welding 0.078-in. steel, heat time 6 cycles, cool time 5 cycles, force 1500 lb and speed 55 ipm

welding to a series of overlapping welds. The welding current is of primary importance in determining the extent of fusion. It thus determines the degree to which the weld metal penetrates the parent metal and the degree to which the individual nuggets overlap for a given current pattern and welding speed. Figures 4, 5 and 7 show the influence of the welding current on nugget penetration and overlap for the three thickness of low carbon steel used at fixed conditions of current pattern, welding speed, and electrode force. As would be expected, increases in welding current cause increases in penetration and overlap of the weld nuggets. For this work with low carbon steel, the following ranges of overlap and penetrations were established as being most desirable; penetration 30–70% and overlap 0–20% with 10–15% considered optimum. As mentioned previously much use was made of the hydrostatic pillow test as a means of studying qualitatively weld strength. Figures 6 and 8 show the relation of pillow sample failure pressure to welding current. As the welding current is increased the failure pressure increases to some maximum value which is dependent on the thickness of the material and the size of the test specimen. Current values in excess of those giving full strength serve to increase the penetration and overlap, but add nothing to strength as the welds are always sufficiently strong that the pillow samples fail in the parent metal adjacent to the weld. It is important to note that this minimum value of current which will produce full strength welds produces nugget penetration on the order of 30%. This may be seen by comparing Figs. 5 and 6 and 7 and 8 and has been borne out by other data obtained in conjunction with various phases of this work. Figure 5 shows the

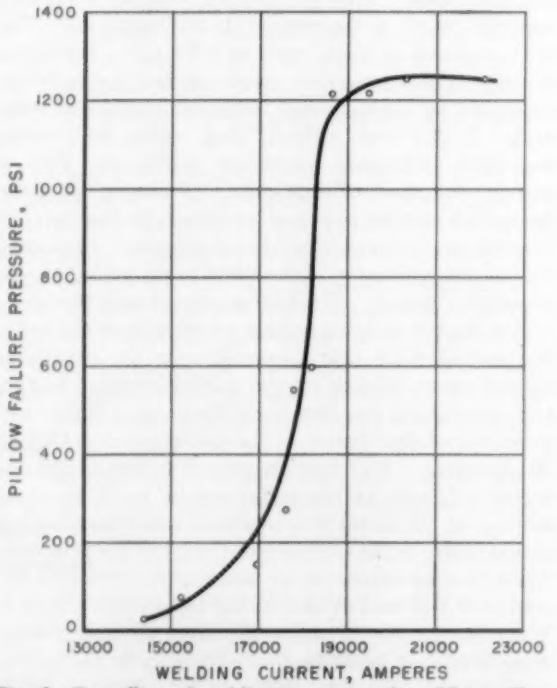

Fig. 8 The effect of welding current on 6- x 10-in. pillow failures of seam welded 0.078-in. steel, heat time 6 cycles, cool time 5 cycles, force 1500 lb and speed 55 ipm

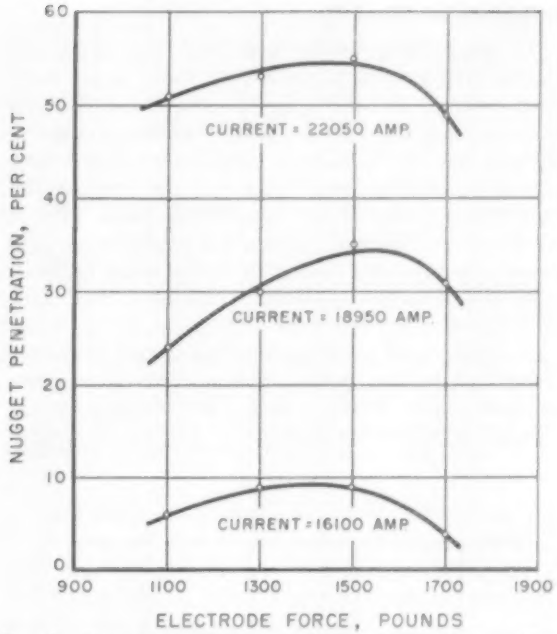

Fig. 9 Nugget penetration in 0.078-in. steel for various electrode forces and currents, heat time 6 cycles, cool time 5 cycles and speed 55 ipm

average nugget properties of 0.049-in. steel as functions of the welding current at the conditions noted. Figure 6 in turn shows the relation of weld strength to welding current. As may be seen welds of full strength correspond approximately to penetration of 30% or greater. Similar remarks apply to Figs. 7 and 8.

Electrode Force

The effects of varying the electrode force upon nugget properties for selected current values are shown in Figs. 9 and 10. Referring to Fig. 9, it can be seen that as the force is increased up to a certain point there is also an increase in nugget penetration. This appears to be reasonable, as with increased electrode force

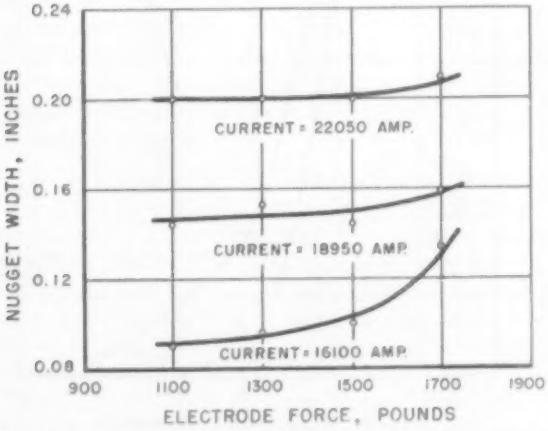

Fig. 10 The effect of electrode force and welding current on nugget width in welding 0.078-in. steel, heat time 6 cycles, cool time 5 cycles, and speed 55 ipm

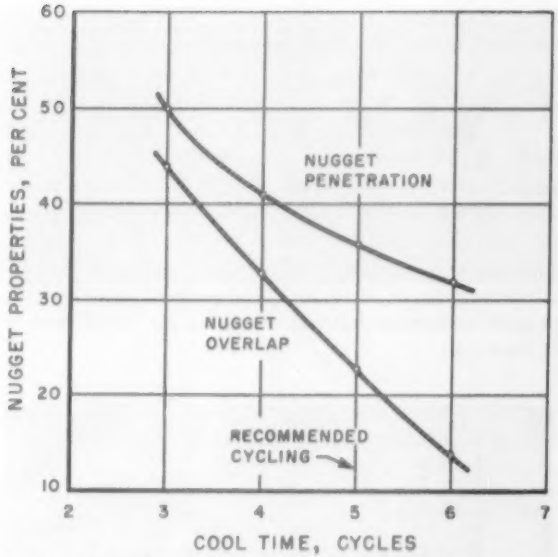

Fig. 11 Nugget penetration and overlap in 0.078-in. steel as effected by varying the cooling time, heat time 6 cycles, force 1500 lb, weld current 18,950 amp and speed 55 ipm

the current is more concentrated at the center due to the crowned electrode surfaces. The point, where the penetration starts decreasing in value, corresponds to the point where the nugget width is increasing as indicated in Fig. 10. Due to the higher forces in this area the sheets are flattened out and the current distributed over a wider space. Hence with a wider nugget width there is a decrease in the nugget penetration. No special tests were made as to the

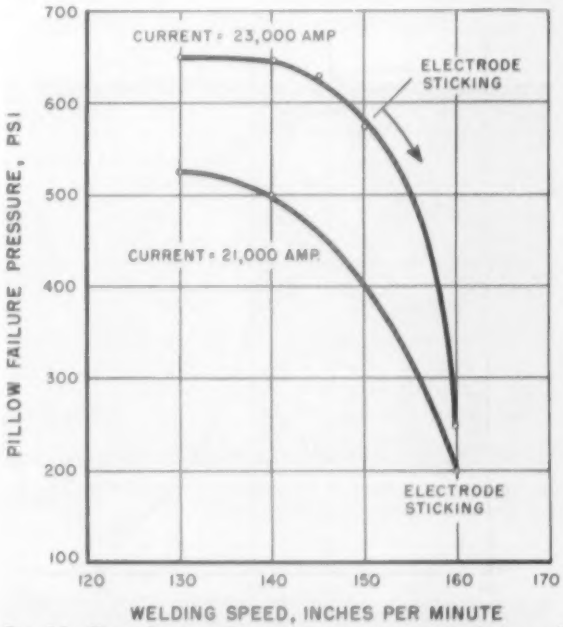

Fig. 12 The influence of welding speed upon the failure of pillow specimen of 0.049-in. steel, heat time 1 cycle, cool time 1 cycle and force 1050 lb

effect of electrode force upon pillow specimen strength, however, it was observed that specimens having penetration values ranging from 30 to 60% gave good strength.

Both figures indicate tests made on 0.078-in. steel and, for this thickness, it was observed that welds of good properties result from the use of a 1500 lb force. Tests on 0.031- and 0.039-in. steels show similar trends as indicated by the above curves. For 0.031-in. sheet a force of 700 lb is recommended and for the 0.049 a force of 1050 lb. These check closely with present recommended values given in the WELDING HANDBOOK.

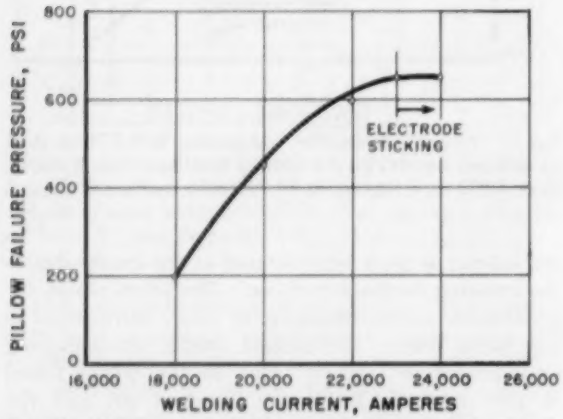

Fig. 13 Effect of welding current on pillow specimen failures for 0.049-in. steel; heat time 1 cycle, cool time 1 cycle, force 1050 lb, and a speed of 140 ipm

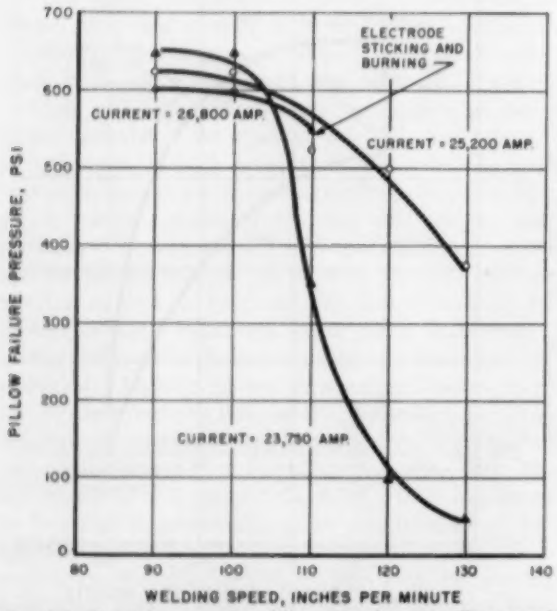

Fig. 14 The influence of welding speed upon pillow specimen failure for 0.078-in. steel; heat time 2 cycles, cool time 1 cycle, and force 1500 lb

Influence of Weld Time

In determining proper heat and cool times the recommended current, speed and force values were held constant and the cycle times were varied over a wide range. From these tests, specimens were examined and optimum cycle times were selected from specimens showing no porosity and with overlap and penetration values within established limits. Figure 11 shows the effect of varying the cooling cycle time for a selected heat time of 6 cycles using 0.078-in. steel. Metallurgical specimens, having a heat time of 6 cycles and a cool time of 5 cycles, indicate that such welds are of good quality with nugget properties within established limits. Similar tests were made on 0.031- and 0.049-in. steel. For average recommended speeds it was determined that a heat time of

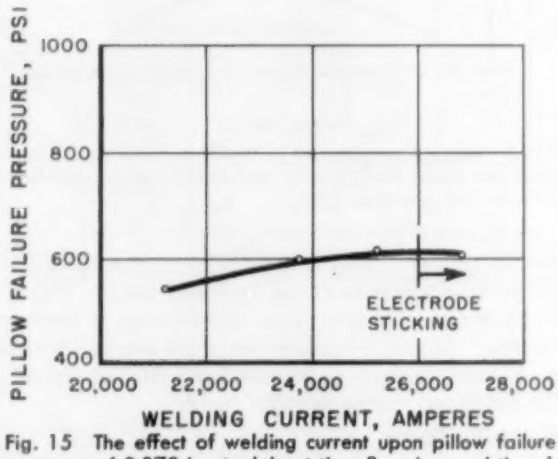

Fig. 15 The effect of welding current upon pillow failure pressures of 0.078-in. steel, heat time 2 cycles, cool time 1 cycle, force 1500 lb, and a speed of 100 ipm

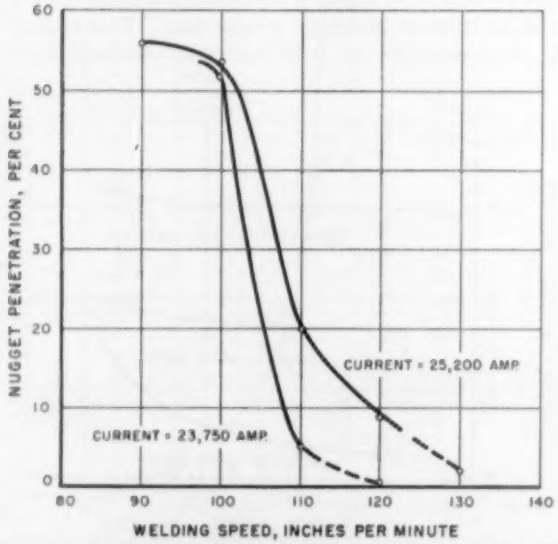

Fig. 16 Effect of welding speed upon nugget penetration in 0.078-in. steel having a heat time of 2 cycles, cool time 1 cycle, and 1500 lb electrode force

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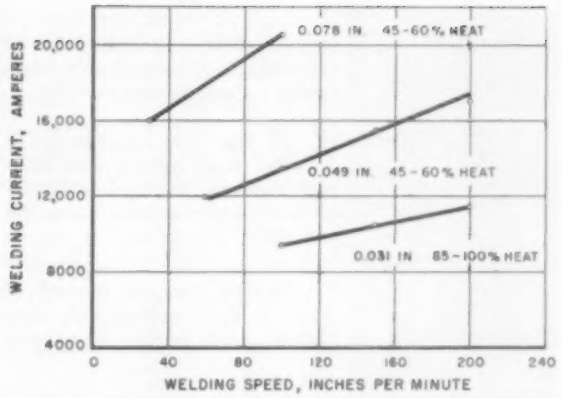


Fig. 17 The influence of continuous welding current upon the welding speed of different thicknesses of sheet steel

4 cycles and a cool time of 3 cycles was satisfactory for 0.049-in. steel. For 0.031-in. steel the values are heat time 3 cycles and cool time 2 cycles. In all cases the nugget size is controlled principally by the heat time and the overlap by the cool time.

Inspection and Testing of Welds

Besides the much used pillow test to determine relative weld strength and leak tightness several metallurgical and radiographic inspections were made. Suitable coupons of the welds were X-rayed with a Westinghouse 150-kv Industrial X-ray Machine to inspect the welds for internal defects and porosity. Typical welds with porosity are shown in Figs. 23 and 24. Welds with little or no porosity are illustrated in Figs. 25 and 26.

The metallurgical inspections were made on both longitudinal and transverse sections through the welds. Several typical longitudinal sections at low magnification are shown in Figs. 18 through 22. The actual determination of nugget penetration and overlap was made at higher magnification with a Bausch and Lomb toolmakers microscope. This made possible a close study of the weld microstructure and accurate measurement of the structures with the micrometer measuring feature of the microscope. A typical study of the characteristic dendridic structure of the welds is shown in Fig. 20. This also shows a portion of the overlap of adjacent nuggets.

Welding Speeds

The speeds at which low-carbon steel may be seam welded leaktight depend upon the nature of the assembly being welded, the thickness of the material, its surface condition and the quality of weld desired. The Conclusions and Recommended Welding Schedules presented at the end of this paper and detailed in Table, 1 were developed from a series of welding schedules on each of the three thicknesses.

A portion of this project was to check the Recom-

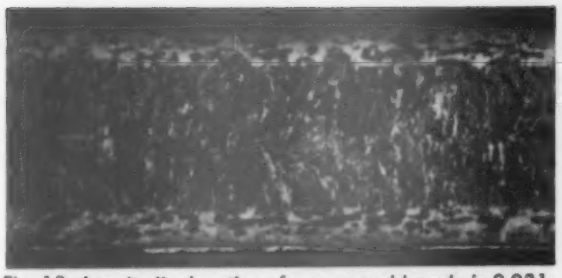


Fig. 18 Longitudinal section of a seam weld made in 0.031-in. low-carbon steel at recommended conditions for 45 ipm. × 20

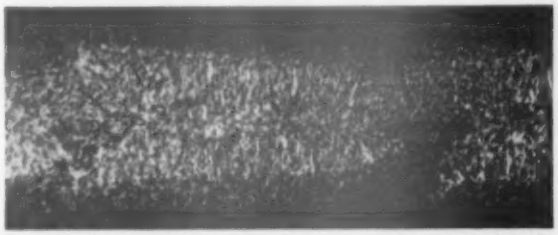


Fig. 19 Longitudinal section of a seam weld made in 0.049-in. low-carbon steel to illustrate the general characteristics of individual nuggets resulting from too long cool time. × 11

Fig. 20 Longitudinal section of a seam weld made in 0.031-in. steel showing the characteristic dendritic structure. × 14

Fig. 21 Longitudinal section of a seam weld made in 0.049-in. low-carbon steel at recommended conditions for 37 ipm. × 20

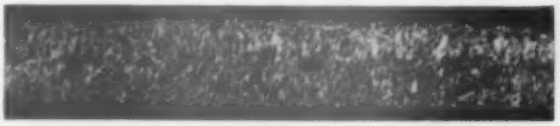


Fig. 22 Longitudinal section of a seam weld made in 0.031-in. steel at recommended conditions for 145 ipm but with higher current of 18,000 amp. Note porosity. × 18

ipm was determined as the maximum satisfactory for

from the Recommended Practices of the RWMA

mended Practices of the RWMA Handbook and accordingly some of the schedules presented were developed about the recommended speeds. For welding with interrupted current at the maximum possible speeds the procedures were modified such that heat and cool times of the shortest usable duration were used and proper speeds then determined. Typical pillow sample and nugget penetration data used for these determinations are presented in Figs. 12 through 16. As may be seen in Figs. 12 and 14 the welds suffer a rapid reduction in strength above certain speeds, even with the use of higher currents. The amount of current which could be used was limited by surface burning and electrode pickup. Using the maximum currents possible the welds were made over a range of speeds and the relative strengths tested by pillow specimen. The maximum speed at which full-strength welds could be made was thus determined. The proper welding current was then determined at those speeds. Current determination data are shown in Figs. 13 and 15 of pillow failure pressure vs. current. Metallurgical data on penetration was also used. A typical curve of penetration vs. welding speed is shown in Fig. 16. All data are for welding with full wave current of 85 to 100% heat on electronic seam welder controls.

In determining schedules for welding with continuous current (Fig. 17), the previously established electrode forces were used thus leaving only the determination of proper current magnitude and waveform to be done. The principal difficulty encountered at the higher speeds was severe surface burning of the metal with the resultant electrode pickup. All schedules presented are designed to avoid this problem and give maximum weld strength. The seam welder used had a maximum speed of 200 ipm. This speed was readily attained with 0.031- and 0.049-in. material but a speed of 100

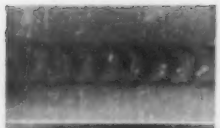


Fig. 23 Print from radiograph of 0.049-in. steel welded with interrupted current showing porosity

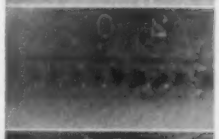


Fig. 24 Print from a radiograph of 0.049-in. steel welded with interrupted current showing internal defects uniformly spaced along the nuggets

Fig. 25 Print from a radiograph of 0.078-in. steel welded with interrupted current showing no porosity

Fig. 26 Print from a radiograph of 0.078-in. steel welded with partial wave continuous current showing no porosity

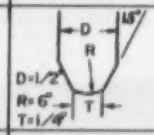
THICKNESS	ELECTRODE SHAPE	MINIMUM CONTACTING SHEET OVERLAP	ELECTRODE FORCE	TIMING CYCLE "ON" CYCLES	TIMING CYCLE "OFF" CYCLES	WELDING SPEED	WELD SPACING W/IN.	WELDING CURRENT AMPS.	WELD OVERLAP	WELD PENETRATION
0.031"		1/2"	700 LB.	2	3	45 1/2	14	12500	21%	52%
				3	2	72	10	13000	18	55
				2	1	103	115	15000	17	34
				1	1	145	124	16500	0	45
0.049"	D=1/2" R=6" T=5/16"	9/16"	1050 LB.	4	4	37	122	14000	30%	50%
				4	3	65	8	16500	25	50
				2	2	95	95	20000	18	35
				1	1	140	13	22500	0	30
0.078"	D=1/2" R=6" T=3/8"	11/16"	1500 LB.	6	6	30	10	16000	20%	50%
				6	5	55	6	20000	30	50
				3	1	85	105	22000	25	50
				2	1	100	12	25000	11	50
SEAM WELDING WITH CONTINUOUS CURRENT										
0.031"	D=1/2" R=6" T=1/4"	1/2"	700 LB.	85–100% OF FULL WAVE		100 1/2		9500		45%
						150		10500		35
						200		11400		30
0.049"	D=1/2" R=6" T=5/16"	9/16"	1050 LB	45–60% OF FULL WAVE		60		12000		30%
						100		13500		32
						150		15500		30
						200		16500		29
0.078"	D=1/2" R=6" T=3/8"	11/16"	1500 LB.	45–60% OF FULL WAVE		30		16000		55%
						100		20500		45

Table 1—Seam Welding Low-Carbon Steel with Interrupted and Continuous Current

ipm was determined as the maximum satisfactory for 0.078-in. material. As welding speed is increased it is necessary to increase the welding current in order to develop sufficient heat in the material. This increased current leads to the aforementioned surface heating problems which limit the speed at which material may be welded, even with flood cooling.

It was found advantageous to use partial wave current on the 0.049 and 0.078 in. thicknesses (Fig. 17). Partial wave current was limited to waveforms corresponding to 45 to 60% heat on a standard 20 to 100% heat electronic seam welder control. Although these reduced waveforms present power factor difficulties their use is dictated if higher speed continuous current welding is to be done satisfactorily.

Conclusions

The summarized results of this work on low-carbon steel are presented in Table 1. They include speeds from the Recommended Practices of the RWMA Handbook and from data prepared by the Taylor-Winfield Co. for the higher speeds. Certain changes were deemed necessary in the various schedules published to date. Those values believed most desirable are presented. It must be noted that the highest speed for interrupted current welding of each of the thicknesses give welds with essentially zero nugget overlap. This is not an optimum condition but the corresponding currents and electrode forces will produce welds with satisfactory strength and pressure tightness as shown in Figs. 12 through 15 so long as the recommended speeds are not exceeded.

The recommended schedules for continuous current are based on use of waveforms as specified for each thickness. It will be found necessary to use electrode forces near the recommended values as severe porosity and burning tend to occur with reduction in the force.

APPLICABILITY OF CHARPY V TESTS

DISCUSSION BY A. AUDIGE AND AUTHOR'S REPLY

The relation between U-notch and V-notch tests, for so long discussed, has been recently examined in France by the IRCN (French Shipbuilding Research Institute) with the help of the French mills and the IRSID (French Iron and Steel Institute).

All the previous archives in France were expressed with U-notch specimens. I wish to mention that the "resilience" specimen with the U notch is of French origin and the keyhole specimens, after Charpy's studies, became the French standard for 30 years.

The new tests with V notch carried out on 26 already known steels gave some support to the correlations previously published and especially with the correlation of Fig. 16 given on the basis of the mid-span temperature generally about 12 ft-lb, or 3.4 kgm/cm.^{2}

There is a close agreement between the French results and the average results of the NSB, the ABS C and A Steels of Fig. 16 of M. Pellini's Paper.

We do not know the origin of the relation formula criticized by the authors: $TTV10 = TT$ mid-span $-$ 20° F. This formula is wrong and there is no straight line representing the relation, both for A steel and for C steels. For A steels, the line proposed by Mr. Gensamer is statistically a good line. For C steels the shift is much greater than for A steels (about 50° F) as indicated by the graph.

We therefore consider Mr. Pellini's criticism is against the relations more than against the U-notch specimen.

On the basis of the French tests, the best relation seems to be given by the curve:

$tU\ 10 = tV\ 10 - 20°$ C for rimmed steels
$tU\ 10 = tV\ 10 - 23°$ C for semikilled steels
$tU\ 10 = tV\ 10 - 30°$ C for killed steels

tU 10 and tV 10 being the transition temperature relating to a 10 ft-lb energy.

It should be remembered that a correlation of this kind could only be of statistical value. For separated steels it might be an appreciable scatter with 6° C as standard deviation. But for a general classification or the choice of a valuable standard, this relation experimentally defined seems to be satisfactory.

The random tests on heats of various countries by the ABS seem to agree with these conclusions. But, unfortunately, the keyhole TT has been related often to 20 ft-lb.

This selected value generally correlates with the upper part of the transition curve almost horizontal; an empirical value of 11 ft-lb (3 kgm/cm^2) on keyhole specimens, selected in France for 4 years, falls into the transition keyhole range, very near to the mid-span temperature of this transition range, which is the most interesting characteristic of the keyhole specimen; its value is a physical one, as being related with the probability of brittle or ductile behavior of the specimen during the test.

(*Authors' Reply on page 156-s*)

A. Audige is associated with the Institut de Recherches de la Construction Navale, Paris and chairman of the French Weldability Committee, Institut de Soudere, Paris.

Paper by Pusak, Schuster and Pellini was published in THE WELDING JOURNAL, 33 (9), Research Suppl., 433-s to 441-s (1954).

THE FLASH WELDING OF COMMERCIAL MOLYBDENUM

Part I: Temperature distribution during parabolic flashing of 1/2-in. sintered-and-wrought molybdenum rods

BY ERNEST F. NIPPES AND WEN H. CHANG

Abstract

A study has been made of the effects of clamping distance and platen acceleration on the temperature distribution during parabolic flashing of 1/2-in. diam sintered-and-wrought molybdenum rods.

The experimental procedure of measuring the temperatures by means of thermocouples is described. The resulting temperature distributions are interpreted on the basis of the thermal conditions prevailing in the specimens as influenced by clamping distance and platen acceleration. The experimental results are compared with, and, in general, substantiated by the findings of a microscopic examination of the as-flashed specimens.

It is concluded that the parabolic flashing of molybdenum consists of two stages divided approximately at the critical burn-off of 0.2 in./specimen. In the first stage, the temperature of a point a given distance from the flashing interface increases with increasing burn-off, whereas in the second stage the point assumes a constant temperature, resulting in a stabilized temperature distribution along the specimen. The gradient of this stabilized temperature distribution, which governs the width as well as the plasticity of the recrystallized zone, is steeper the shorter the clamping distance and the higher the platen acceleration.

Introduction

Room temperature embrittlement of recrystallized molybdenum is known as one of the major factors which limit the industrial usefulness of the metal. Commercial molybdenum, either sintered or arc cast, exhibits considerable ductility in the as-wrought condition but becomes brittle after it has recrystallized in the presence of even minute amounts of oxygen. In conventional welding processes, the combined effect of high temperature and oxidizing atmosphere is usually so strong that the weld joints are almost always weak and brittle when tested at room temperature. A survey of the literature indicates that high-purity molybdenum, obtained only by elaborate means, may not be essential to the production of welds of acceptable ductility; the primary requirement lies rather in the choice of a welding process capable of controlling the absorption, retention and distribution of oxygen. Such a process should be rapid acting to avoid severe oxidation and should also be capable of applying pressure in making the joint. The latter is important not only in plastically deforming the weld structure but also in eliminating the oxidized material near the joint made in a contaminating atmosphere.

It is for this reason that the present research of flash welding of molybdenum was undertaken. By nature of this process, the investigation was divided into two parts: Part I, temperature distribution during parabolic flashing of sintered molybdenum and Part II, production and testing of molybdenum flash welds. The present paper is concerned with the results of Part I. The results of Part II will be presented later.

Literature Review

Molybdenum has been welded by the gas,[1] spot,[2, 3] seam,[3] percussion,[3] inert-gas arc,[1, 3-5] submerged arc[4] and atomic hydrogen[1, 5] processes. With the exception of the spot welds made by Kiefer,[2] which were described as ductile, the welds produced by these processes were characterized by low strength and brittleness, presumably resulting from intergranular oxidation, grain coarsening and porosity in the heat-affected zone. No information of flash welding of molybdenum seems to have been reported in the literature except that, in 1948, Heuschkel[3] stated that molybdenum could not be flash welded because of the formation of volatile molybdenum oxide.

In recent years, workers at the Battelle Memorial Institute, through their intensive research on the cause of molybdenum embrittlement, have reported the following significant findings:[6]

1. Multiple high-vacuum (less than 0.1 μ) arc-cast molybdenum was ductile in the as-cast state.

2. Commercial sintered molybdenum rods "solid-state purified," i.e., vacuum-annealed at 3500–4000° F for 12–24 hr, had considerably lower transition temperatures (−50 to −90° F at a strain rate of 0.064 per second) than in the as-received condition (5° F), as evaluated by a bend test. This improvement of ductility was attributed to the removal of oxygen and nitrogen from the metal. Its significance lies in the fact that recrystallization *per se* may not be a fundamental cause of the room-temperature embrittlement of molybdenum.

3. Ductile upset welds could not be made of the "solid-state purified" molybdenum in dry hydrogen or helium atmosphere, but could be made in high vacuum.

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Report prepared for Metallurgical Division, Oak Ridge National Laboratory, Oak Ridge, Tenn., operated by Carbide and Carbon Chemical Co., a division of Union Carbide and Carbon Corp.

4. Butt welds made of commercial wrought sintered molybdenum or wrought arc-cast molybdenum, in dry hydrogen or helium, were very brittle. The ductility could not be improved by postweld heat treatments.

The impotence of gaseous atmospheres in molybdenum welding was also demonstrated in the research conducted at the Massachusetts Institute of Technology. It was reported[5,7] that butt welds made in purified hydrogen, helium, chlorine, or carbon tetrachloride atmospheres were brittle and low in strength. However, somewhat better welds could be made in carbon tetrachloride, under carefully controlled heating conditions, by inserting a sodium-lithium metal fill in the joint.

The fact that commercial molybdenum, either sintered or arc cast, has considerable ductility in the worked condition indicates that the distribution as well as the amount of oxygen are related in the embrittlement of the metal. J. Wulff[7] believes that ductile welds can be made if the weldment is subjected to sufficient plastic deformation to break up and redistribute the grain boundary oxide-film formed during welding.

More recently Moss[8] reported success in producing satisfactory thin-sheet (0.5 mm) molybdenum welds by impact welding in a purified hydrogen or argon atmosphere. His work was characterized by plastic deformation of the joint at low temperatures (690–1400° C) at which no appreciable recrystallization or grain growth took place. By chromium-plating the faying surfaces or inserting foils of different materials, such as austenitic stainless steel, tantalum, niobium, etc., the amount of deformation necessary to effect the welding could be materially reduced.

Purpose and Scope

The purpose of this part of the research was to investigate the nature and extent of the influence of the variables which affect the temperature distribution established in molybdenum specimens during the flashing cycle of the flash-welding process. This was considered necessary because the properties of molybdenum welds made in a contaminating atmosphere depend on the extent to which recrystallization has taken place. The latter is primarily a function of the temperature to which the material is heated and the time it is exposed to these temperatures.

In flash welding, heat is produced mainly by the flashing arc which consumes the material at the interface and partly by the Joule (I^2Rt) effect of the current flowing through the entire work being welded. For a given material, the rate of energy input will be some function of the rate of burn-off which, in turn, is controlled by the flashing pattern, i.e., the pattern of the platen movement. Two patterns are in commercial use: linear flashing in which the platen moves with a constant velocity, and parabolic flashing in which the platen moves with a constant acceleration. In a previous investigation of the flash welding of steel, conducted at Rensselaer Polytechnic Institute,[9] it was found that less burn-off was required in parabolic than in linear flashing before a stable temperature distribution was established. Consequently, the parabolic flashing pattern was used exclusively in the present study.

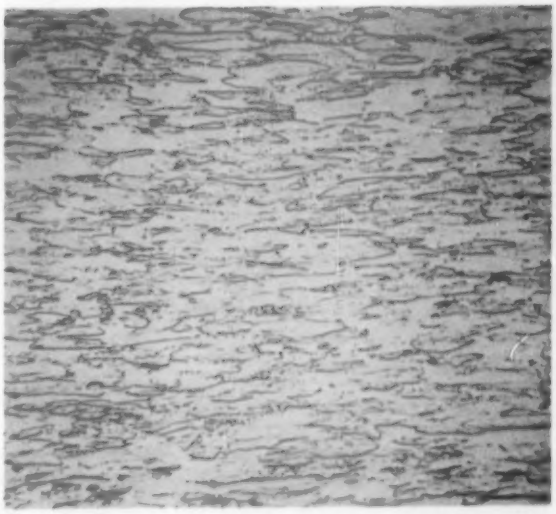

Fig. 1 Photomicrograph of as-received molybdenum

When heat is being produced by the flashing arc and the Joule effect, it is also constantly conducted away by the water-cooled copper clamping jaws which hold the two specimens in position. The rate of cooling will be governed by the clamping distance (the length of the unclamped parts of the work between the jaws) and the shape and thickness of the work.

For a given flashing pattern, therefore, the important variables which influence the temperature distribution are amount of burn-off, platen acceleration, clamping distance, and dimensions of the work. The effects of these variables, except the last one which could not be included because of limited stock supply, constituted the main scope of the present investigation. To correlate the experimentally determined temperature distributions with the corresponding microstructural changes, this investigation was also accompanied by a microscopic study of the as-flashed specimens.

Experimental Work

Material

The study of temperature distribution was confined to $^1/_4$-in. sintered and wrought round rods manufactured by the Fansteel Metallurgical Corp. This material contained 0.0049% carbon, 0.0015% oxygen, and less than 0.0001% nitrogen. The fibered structure of the as-received molybdenum is shown in Fig. 1.

Apparatus

The flash welding was performed on a 300-kva, hydraulically operated, Federal Flash Welder in the Welding Laboratory, Rensselaer Polytechnic Institute. This welder, powered from a single-phase alternator driven by a 3-phase synchronous motor, had been modified to be used for either linear or parabolic flashing, or both.[10] With respect to parabolic flashing, it was provided with the following features:

1. A method of initiating the chosen flashing pattern by the first pulse of the flashing current.
2. An adjustable constant velocity of approach during the interval preceding the initial specimen-to-specimen contact.
3. Calibrated control dial for selecting the platen acceleration.

In conjunction with the flash welder, a 14-channel Consolidated recording oscillograph was used. The recording galvanometers had a sensitivity of 0.03 ma/in. at the recording distance.

Experimental Procedure

Flashing Operation. In accordance with the purpose of this work, five platen accelerations, 0.006, 0.024, 0.060, 0.090 and 0.120 in./sec^2 were chosen. For each platen acceleration, flashing was accomplished at three clamping distances of 1.5, 2.0 and 2.5 in. At least four runs were made for each combination of platen acceleration and clamping distance. A total platen travel, or burn-off, of 0.8 in. was selected for convenience.

The specimen size was initially $^1/_2$ x 6 in. but was later changed to $^1/_2$ x 4 in. for material economy reasons. The specimen preparation consisted of machining the end faces in order to obtain uniform flashing over the abutting surfaces. Four or five 0.052-in. thermocouple holes, of increasing depth and all inclined 45 deg to the specimen axis, were drilled in one of the pair of specimens to be flashed. This thermocouple-hole arrangement is shown in Fig. 2. The exact location of each hole, and hence the distance of each thermocouple from the initial interface, was revealed by radiographing the specimen.

In making a run, the specimens were clamped in the water-cooled jaws which had been set at a predetermined clamping distance. By means of a condenser-discharge arrangement, the thermocouples were welded to the bottom of the holes. The selection of thermocouple material was one of the major difficulties encountered in this phase of the work because of the very high temperatures and the violent flashing are present near the interface. The only high-temperature thermocouple available in fine-wire form, was that of Pt−Pt+10%Rh. This was used in the first one or two holes nearest the interface. For the remaining holes, chromel-alumel thermocouples were employed. Attempts were made to drill the holes farther back from the end of the specimen but failed to prolong the life of the Pt−Pt+10%Rh thermocouple materially so that a sufficiently long record could be obtained.

Due to its irregular wave shape and intermittent as well as transient nature, it was not possible to measure the flashing current. The flashing voltage was of the order of 15 v. According to a previous study of the flashing of aluminium[11] and steel,[9] the flashing voltage was found to have little effect on the temperature distribution. During flashing, the output of the thermocouples and the signal of platen travel were fed to separate galvanometers on the oscillograph. One of such records is shown in Fig. 3.

Microscopic Examination of Flashed Specimens. As a result of the flashing, certain microstructural changes such as formation of oxide and porosity, recrystallization, and grain growth, take place in the molybdenum. Some of these changes, for example, recrystallization and grain growth, are known to be temperature-time dependent. A microscopic study was therefore made of the flashed specimens in order to correlate these changes with the temperature distribution obtained under various platen accelerations and clamping distances. All specimens were manually polished and etched with Murakami's reagent.

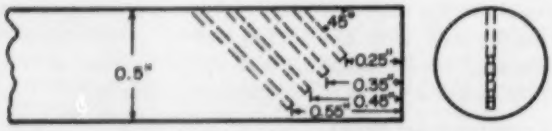

Fig. 2 Sketch showing thermocouple arrangement

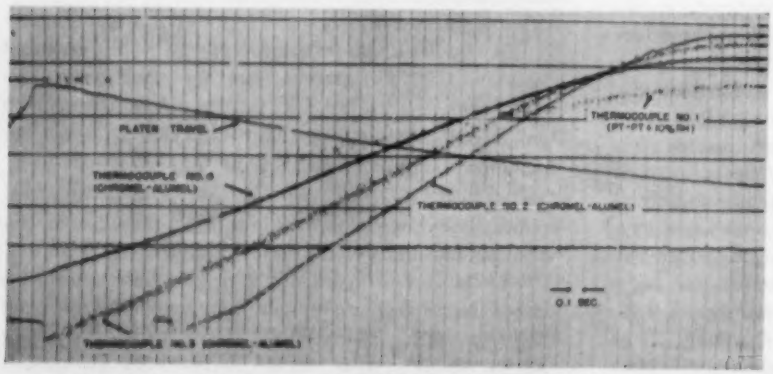

Fig. 3 Photograph of an oscillographic record of a typical flashing cycle

Data, Results and Discussion

1. Methods of Presenting Results

A total of 15 groups of specimens were flashed, each group corresponding to a given combination of platen acceleration and clamping distance. The results are summarized in Table 1. These results were derived from the experimental data through the following treatment. The galvanometer deflections on each oscillographic record were first converted respectively into temperatures and total burn-offs (platen travel) by means of calibration curves. These temperatures were the instantaneous temperatures experienced by the thermocouples and hence by the four or five fixed points (locations of the thermocouples) whose distances from the instantaneous interface decreased continuously with the progress of flashing. For reasons that will im-

Table 1—Temperature Distribution During Parabolic Flashing of $^1/_2$-In. Sintered and Wrought Molybdenum Rods

Group	Platen acceleration, in./sec²	Clamping distance, in.	Distance from instantaneous interface, in.	Stable temperature, °F
1	0.006	2.5	0.35	2800
			0.40	2600
			0.45	2400
			0.50	2200
			0.55	2000
2	0.006	2.0	0.30	3150
			0.35	2750
			0.40	2550
			0.45	2350
			0.50	2150
			0.55	1950
			0.60	1750
3	0.006	1.5	0.30	2750
			0.35	2450
			0.40	2250
			0.45	2050
			0.50	1850
			0.55	1650
4	0.024	2.5	0.25	3000
			0.30	2600
			0.35	2300
			0.40	2100
			0.45	1900
			0.50	1650
5	0.024	2.0	0.20	3300
			0.25	2900
			0.30	2550
			0.35	2250
			0.40	2000
6	0.024	1.5	0.25	2750
			0.30	2450
			0.35	2150
			0.40	1900
			0.45	1650
			0.50	1450
7	0.060	2.5	0.25	2750
			0.30	2350
			0.35	2000
			0.40	1750
			0.45	1550
			0.50	1350
8	0.060	2.0	0.15	3400
			0.20	3050
			0.25	2550
			0.30	2200
			0.35	1800
9	0.060	1.5	0.10	3600
			0.15	3300
			0.20	2850
			0.25	2500
			0.30	2100
			0.35	1750
10	0.090	2.5	0.15	3350
			0.20	2850
			0.25	2450
			0.30	2050
11	0.090	2.0	0.15	3250
			0.20	2750
			0.25	2400
			0.30	2050
			0.35	1700
12	0.90	1.5	0.15	3200
			0.20	2650
			0.25	2250
			0.30	1900
			0.35	1550
			0.40	1350
			0.45	1100
13	0.120	2.5	0.15	3100
			0.20	2700
			0.25	2300
			0.30	1900
			0.35	1500
14	0.120	2.0	0.10	3300
			0.15	2950
			0.20	2450
			0.25	2200
			0.30	1850
			0.35	1400
15	0.120	1.5	0.15	3000
			0.20	2450
			0.25	2100
			0.30	1700
			0.35	1350

mediately be made clear, it was more convenient to plot these temperatures as a function of the burn-off per specimen (assumed to be one half of the total burn-off) rather than of the time. From each oscillographic record, a family of curves, one for each thermocouple, was thus produced, similar to that shown in Fig. 4. The next step was to find the instantaneous temperatures of the points at arbitrarily chosen constant distances relative to the moving flashing interface, i.e., points "moving" with an imaginary instantaneous velocity equal to that of the interface. To do so required a knowledge of the distance of each thermo-

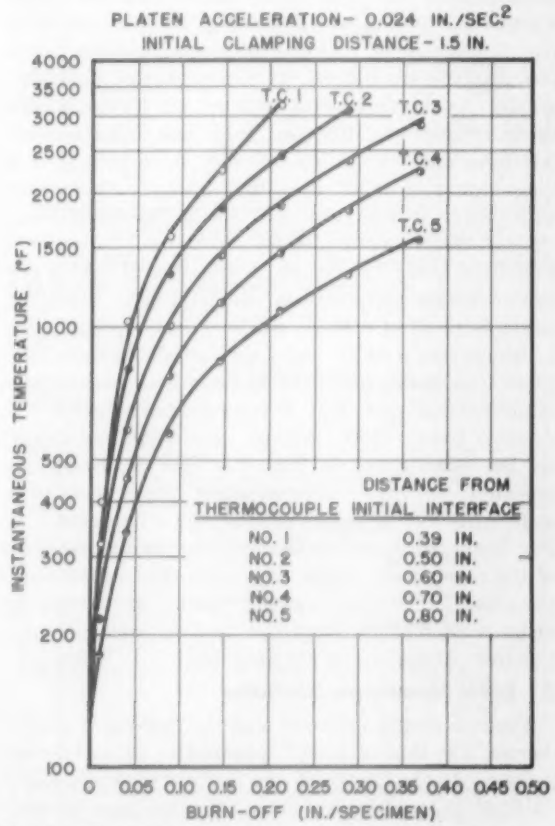

Fig. 4 Effect of burn-off on measured thermocouple temperature

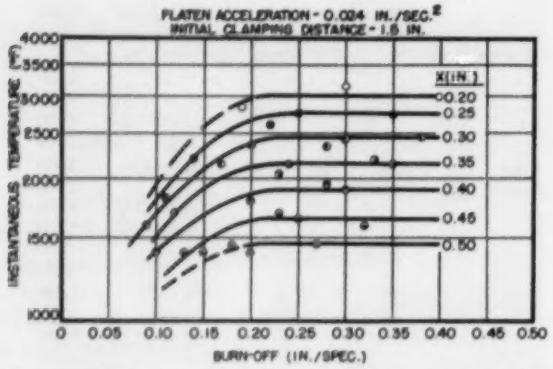

Fig. 5 Effect of burn-off on instantaneous temperature distribution

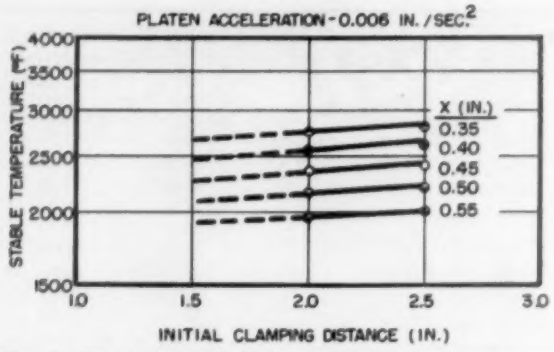

Fig. 6 Effect of initial clamping distance on temperature distribution. Groups 1, 2 and 3

couple from the initial interface, and this was obtained from a radiograph of the specimen made prior to flashing. Using this knowledge, it was then possible to plot, from the records pertaining to each group, the instantaneous temperatures of the "moving" points at constant distances from the flashing interface as a function of burn-off per specimen. An example of such a plot is presented in Fig. 5, the data having been obtained from Fig. 4. The actual plotting may best be understood by an illustration. Let it be desired, for instance, to find the instantaneous temperatures of a "moving" point at a distance of 0.30 in. from the instantaneous interface under the flashing conditions represented by Fig. 4. Beginning with thermocouple No. 1, it is noticed that the thermocouple-initial interface distance was 0.39 in., as measured from the radiograph. At the burn-off of 0.09 in., per specimen, this thermocouple was 0.30 in. from the instantaneous interface and, according to Fig. 4, experienced a temperature of 1600° F at that instant. One point, with coordinates of 1600° and 0.09 in./specimen burn-off, was therefore obtained for the $x = 0.3$-in. curve in Fig. 5, x being the constant distance at which the instantaneous temperatures are required. Similarly, at the burn-off of 0.20 in./specimen, thermocouple No. 2, which was 0.50 in. from the initial interface, also came to a distance of 0.30 in. from the instantaneous interface and now had a temperature of 2400° F. Another point (2400°, 0.20 in./specimen) was located on the same curve in Fig. 5. This procedure was repeated for every thermocouple which yielded a sufficiently long record to permit such a treatment. In Fig. 5, the curve drawn through the average positions of the points so obtained represents the instantaneous temperatures of the "moving" point, at a constant distance of 0.30 in. from the flashing interface, as a function of the burn-off per specimen.

2. *Stable Temperature Distribution*

Figure 5 clearly indicates that the flashing of molybdenum, like that of steel,[9] appeared to be a two-stage process. In the first stage, the flashing arc was initially confined to localized regions but became more generalized as the energy input increased with increasing burn-off. This resulted in a rise of the temperatures with the burn-off. Eventually, a critical burn-off was reached at which the interface as a whole attained the melting temperature of molybdenum and remained at that temperature thereafter. This marked the beginning of the second stage, in which heat was generated in an increasing rate as a result of the platen acceleration. However, the cooling effect of the clamping jaws was simultaneously speeded up by the same acceleration with which a given point approached the jaws. A dynamic equilibrium was thus maintained between heating and cooling of the specimen throughout the second stage. Consequently, the temperature of a point at a given distance from the flashing interface became constant and independent of further burn-off. In other words, the specimen assumed a stabilized temperature distribution during the second stage, as indicated by the horizontal portion of the curves in Fig. 5.

The critical burn-off separating the two stages of flashing occurred at approximately 0.2 in./specimen and was not appreciably affected by the flashing conditions and specimen dimensions. Previous investigations on flashing of steel and aluminum alloys[7,9] indicated that the value of the critical burn-off was larger in linear than in parabolic flashing and, in general, increased inversely with the thermal diffusivity of the specimen.

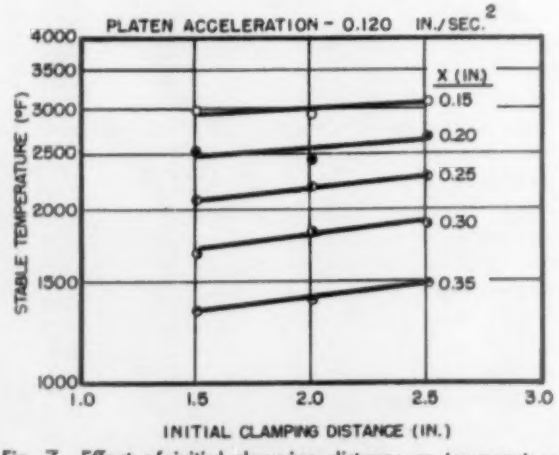

Fig. 7 Effect of initial clamping distance on temperature distribution. Groups 4, 5 and 6

The stable temperature distribution has an important effect on upsetting. As will be shown in Part II, the ductility of the arc-cast molybdenum flash welds was found to be critically influenced by the plasticity of the region adjacent to the flashing interface. In view of the temperature dependence of plasticity, it follows that two welds, otherwise identical, will not possess the same ductility if they differ in temperature distribution at the instant of upsetting. Since variations in burn-off

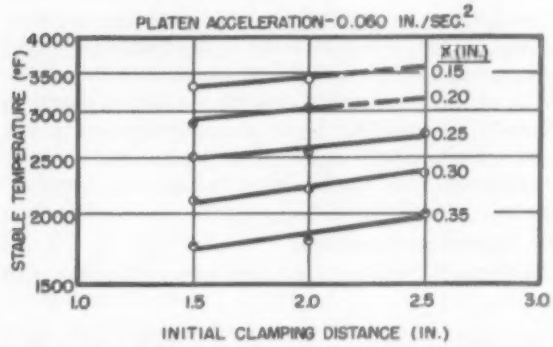

Fig. 8 Effect of initial clamping distance on temperature distribution. Groups 7, 8 and 9

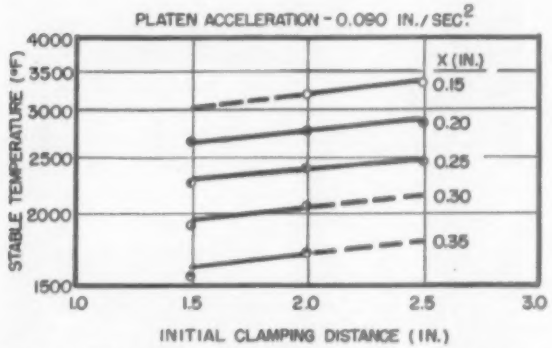

Fig. 9 Effect of initial clamping distance on temperature distribution. Groups 10, 11 and 12

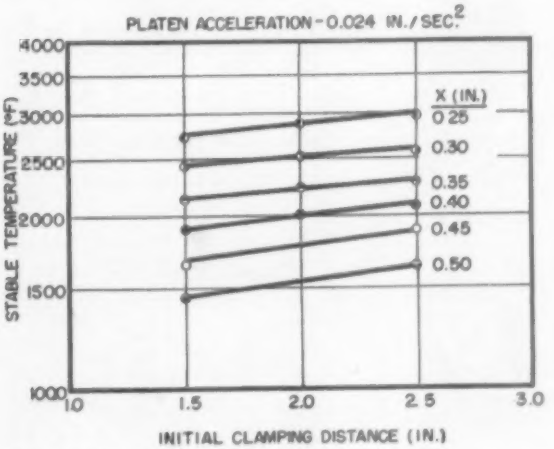

Fig. 10 Effect of initial clamping distance on temperature distribution. Groups 13, 14 and 15

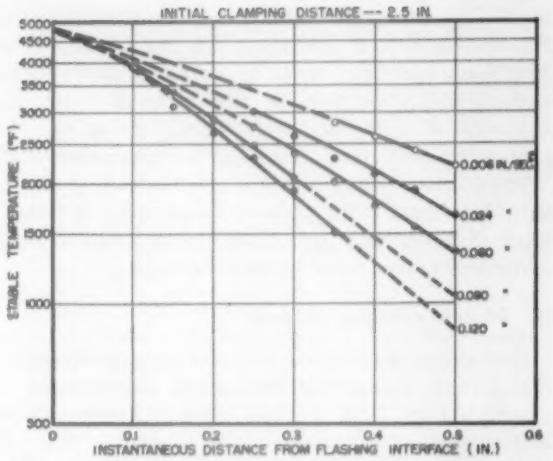

Fig. 11 Effect of platen acceleration on temperature distribution. Initial clamping distance 2.5 in.

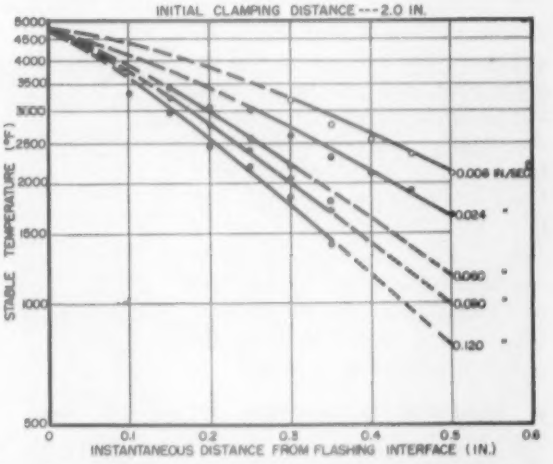

Fig. 12 Effect of platen acceleration on temperature distribution. Initial clamping distance 2.0 in.

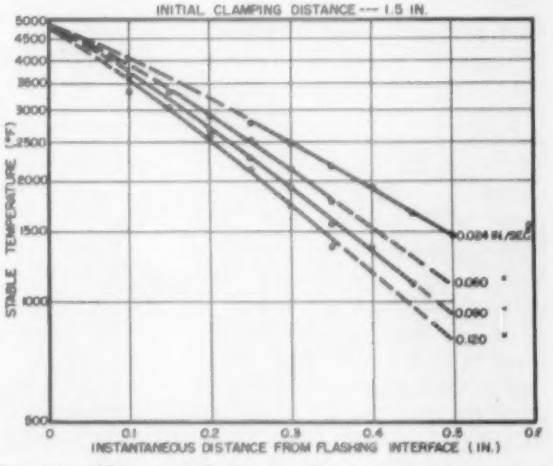

Fig. 13 Effect of platen acceleration on temperature distribution. Initial clamping distance 1.5 in.

beyond the critical burn-off will not change the temperature distribution, it is obvious that the attainment of consistent upsetting under given platen acceleration and clamping distance is greatly assisted by the existence of a stable temperature distribution. In connection with this, it should be mentioned that, whenever economy of material is concerned, upsetting can be initiated with parabolic flash welding after a burn-off of only 0.2 in./specimen, whereas twice as much is required in the case of linear flash welding.[9]

3. *Effect of Clamping Distance*

The effect of clamping distance upon temperature distribution during the flashing of molybdenum is shown in Figs. 6–10, in which the stable temperatures are plotted against clamping distance for different distances (designated as X in the figures) from the flashing interface. The resulting straight lines of these semilogarithmic plots indicate that, other things being equal, the stable temperatures decreased exponentially with decreasing clamping distance. Under conditions wherein the loss of heat due to radiation to the atmosphere was not appreciable, as in the present case, the clamping distance was the predominating factor in determining the length of heat-flow path and, hence, in limiting the rate at which heat could be conducted to the jaws. Consequently, the stable second-stage temperature of a point at a given distance from the flashing interface, would be lower the shorter the clamping distance.

4. *Effect of Platen Acceleration*

In the semilogarithmic plots of Figs. 11, 12 and 13, the stable temperatures are plotted as a function of the instantaneous distance from flashing interface, for various platen accelerations. It is seen that at a given clamping distance, increasing the platen acceleration resulted in a steepened temperature gradient so that the temperature of a given point behind the flashing interface was lower the greater the platen acceleration. This is to be expected in view of the limiting influence imposed upon the rate of heat flow by the thermal diffusivity of the specimen. At high accelerations heat was liberated at a faster rate at the interface than its rate of dissipation along the specimens to the clamps. Thus heat tended to concentrate near the interface, resulting in a steepened temperature gradient.

The consequence of short thermocouple life is evident in Figs. 11, 12 and 13. For slow platen accelerations and long clamping distances, a gentle temperature gradient was produced and temperature measurement was not possible in regions less than 0.3 or 0.35 in. from the interface. For high accelerations and short clamping distances, 0.1 in. was about the limit for accomplishing temperature measurements. Beyond these limits, all the curves had to be extrapolated. The extrapolation is, however, justified by the fact that every curve must terminate at the melting point of molybdenum (taken to be 4760° F), which is the temperature of the interface. Furthermore, in the flashing of steel,[9] the corresponding experimental curves had the simple form as the extrapolated part of the curves shown in Figs. 11, 12 and 13. There was no reason to believe that complications should arise in the temperature distribution near the interface in the case of molybdenum.

5. *Microscopic Examination*

In view of the effects of platen acceleration and

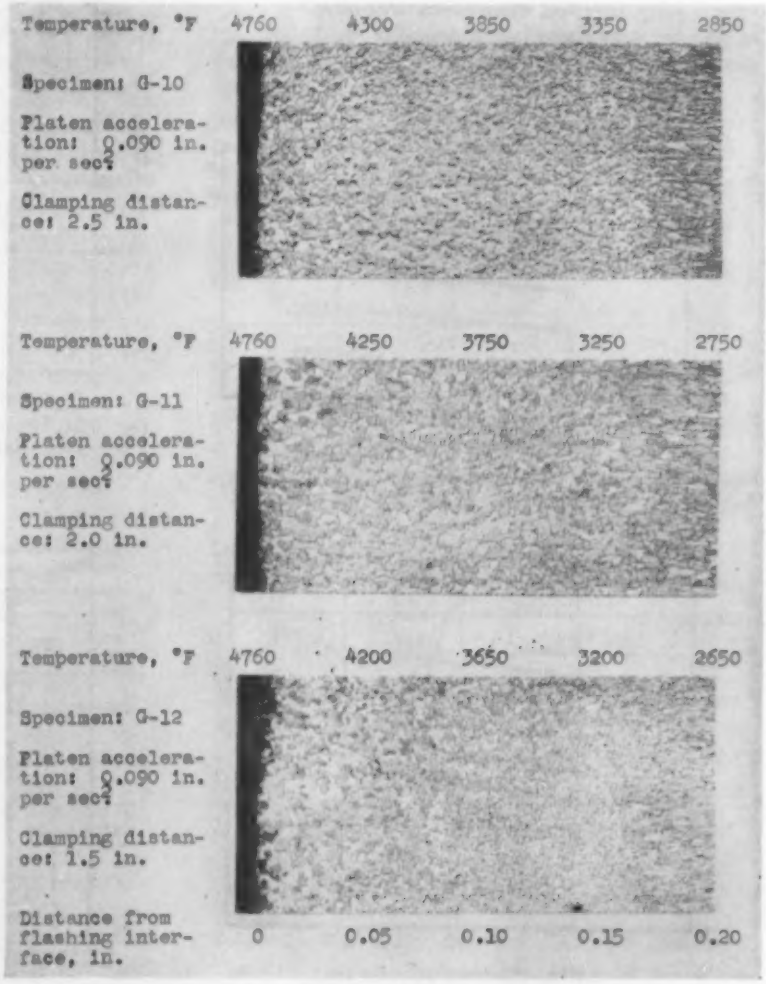

Fig. 14 Photomicrographs showing effect of clamping distance on recrystallized zone. X 14

clamping distance upon the temperature distribution, it is only logical to expect that any temperature-time dependent microstructural changes taking place during flashing will proceed to different degrees under different flashing conditions. The recrystallization zone, for instance, will be wide in a specimen flashed under slow platen accelerations simply because of the long exposure and the gentle temperature gradient. A wide recrystallization zone will also be produced by a long clamping distance which lengthens the heat-flow path and therefore reduces the cooling effect.

The results of the microscopic study, in general, substantiated such correlation between microstructure and temperature distribution. Figure 14 shows the microstructure of three specimens, flashed under the same platen acceleration of 0.090 in./sec^2 but at different clamping distances of 2.5, 2.0 and 1.5 in., respectively. The corresponding temperatures at each distance from the interface were taken, respectively, from Figs. 11, 12 and 13. The effect of clamping distance on the width of recrystallization zone and hence on the temperature distribution, is clearly as predicted and is summarized in Table 2. It is of interest to note that, since the specimens were flashed under the same platen acceleration (0.09 in./sec^2), and total platen travel (0.8 in.), they were exposed to elevated temperatures for an identical length of time (3.4 sec under these conditions). The transition zone between the recrystallized and base metal should therefore occur at different distances from the interface but at about the same temperature, as illustrated in Table 2 which indicates that the transition zone occurred at a temperature of 3200° F.

To illustrate the effect of platen acceleration, the microstructures of an additional set of three specimens are presented in Fig. 15. These specimens were flashed at the same clamping distance of 1.5 in. but under platen accelerations of 0.024, 0.060 and 0.120 in./sec^2, respectively. From Fig. 15, the information given in Table 3 is obtained. The microscopic examination again confirmed the previously established experimental results, namely, slow platen accelerations were conductive to gentle temperature gradients and hence to wide recrystallized zones. In these specimens, although the flashing time was different (6.8, 4.2 and 2.9 sec for G-6, G-9 and G-15, respectively), the fluctuation in the transition-zone tempera-

Table 2—Effect of Clamping Distance on the Width of Recrystallized Zone

		Transition zone	
Specimen	*Clamping distance, in.*	*Distance to interface, in.*	*Temperature, ° F*
G-10	2.5	0.17	3250
G-11	2.0	0.16	3200
G-12	1.5	0.15	3200

Platen acceleration = 0.09 in./sec^2.

Table 3—Effect of Platen Acceleration on the Width of Recrystallized Zone

		Transition zone	
Specimen	*Platen acceleration, in./sec^2*	*Distance to interface, in.*	*Temperature, ° F*
G-6	0.024	0.19	3300
G-9	0.060	0.17	3150
G-15	0.120	0.16	3000

Clamping distance = 1.5 in.

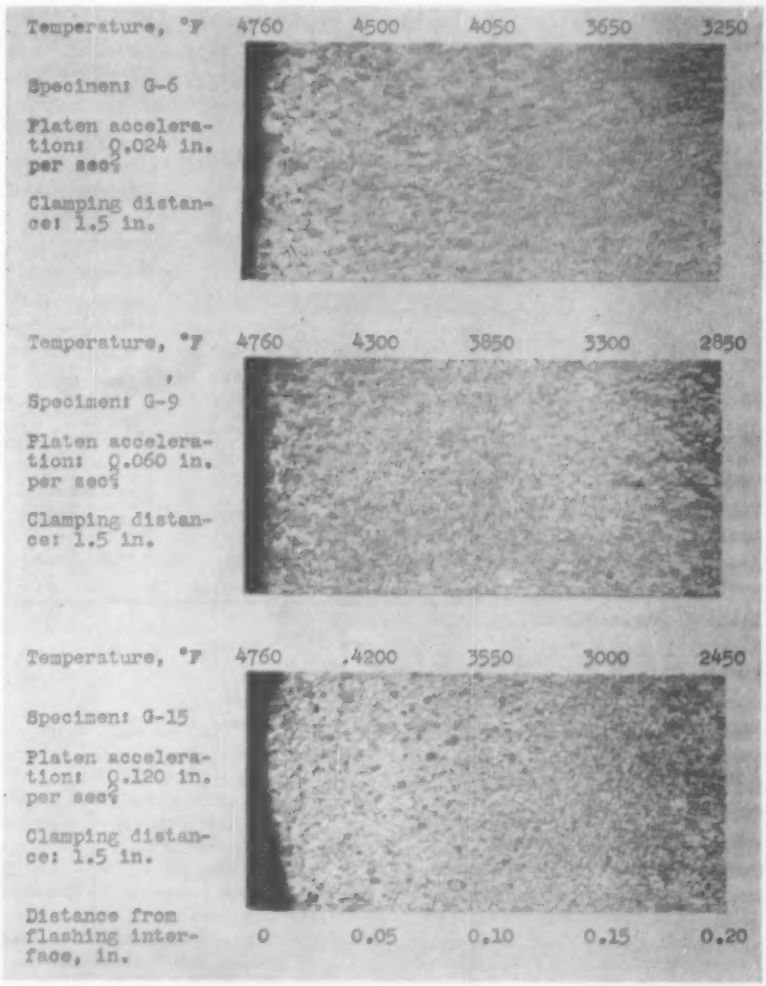

Fig. 15 Photomicrographs showing effect of platen acceleration on recrystallized zone. X 14

tures was not great (3150° ± 150° F). This follows from the fact that recrystallization is not as sensitive to time as it is to temperature. Consequently, when the flashing times were all short, their difference no longer affected the extent of recrystallization significantly. Depending on the flashing conditions, this temperature fell within the range of 0.15 to 0.35 in. for the specimen dimensions used in this investigation.

The microscopic examination showed that oxides, porosity and grain coarsening produced in the molybdenum during the flashing are confined in the recrystallized zone and, in particular, concentrated in the region near the flashing interface. The removal of this region, therefore, will be essential before a ductile joint can be made.

Conclusions

Although the study of temperature distribution was conducted on $^1/_2$-in. sintered molybdenum rods only, it is felt that the results so obtained are nevertheless useful in that they furnish a knowledge of the general effects of the flashing variables upon parabolic flashing of molybdenum. As will be shown in Part II, much of the information derived from the present investigation can be utilized to advantage in understanding the complications encountered in upsetting.

The results obtained in the present investigation have led to the following conclusions:

1. In the parabolic flashing of molybdenum, the flashing cycle was found to be two-staged. In the first stage the instantaneous temperatures increased with burn-off, whereas in the second stage a stabilized temperature distribution existed. Such a stabilized temperature distribution is believed to be essential in securing consistent upsetting.

2. For the parabolic flashing pattern, the second stage began after a total burn-off of 0.4 in., irrespective of other flashing variables. A desired temperature distribution could thus be maintained under appropriate conditions without burning off more than 0.2 in. specimen.

3. Other things being equal, the stable temperatures decreased exponentially with decreasing clamping distance.

4. At a given clamping distance, the temperature gradient became steeper the higher the platen acceleration. A narrow heat-affected zone could therefore be obtained by a high platen acceleration, a short clamping distance, or a combination of both.

5. The effects of platen acceleration and clamping distance, as evaluated by the thermocouple measurement during flashing, were in general agreement with the results of the microscopic study of the as-flashed specimens.

6. Depending on the flashing conditions, the width of the recrystallized zone fell within the range of 0.15 to 0.35 in. This zone was shown to be oxidized and porous and represented the material which should be removed in subsequent upsetting.

Acknowledgments

The authors wish to acknowledge the sponsorship of this project (under Subcontract No. 371 of Contract No. W-7405 eng-26) by the Metallurgical Division of the Oak Ridge National Laboratory, operated by Carbide and Carbon Chemical Co., a division of the Union Carbide and Carbon Corp. They are also indebted to W. D. Manly and P. Patriárca of ORNL for their guidance and suggestions during the course of this investigation.

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RESEARCH NEWS

New Chairman of Pressure Vessel Research Committee

The Pressure Vessel Research Committee of the Welding Research Council announces the selection of F. L. Plummer as Chairman of its Executive and Main Committees, effective Jan. 1, 1955, to complete the unexpired term vacated by the retirement of P. R. Cassidy, who had been Chairman since September 1949.

Mr. Plummer has been closely associated with PVRC for many years and has served as Chairman of the Fabrication Division since 1949. He has also been active in the American Welding Society and represented the Society on several Committees. He was President of AWS for two terms, from 1952 to 1954, and now serves on its Board of Directors as Junior Past-President.

Since 1940, Mr. Plummer has been Chief Engineer and Director of Engineering of the Hammond Iron Works in Warren, Pa. The company fabricates and constructs all types of shop and field-erected steel plate structures, and vessels used by the chemical, rubber and petroleum industries.

Mr. Plummer holds degrees from Ohio University, Oberlin College and Case School of Applied Science, and is licensed to practice Professional Engineering in the states of Ohio, New Jersey and New York. He is author of more than 30 articles published in various technical journals, and has written two books in the structural engineering field.

(*Research News continued on page 152-s*)

TEMPER BRITTLENESS OF PRESSURE VESSEL STEELS

Temper brittleness may contribute to brittle behavior of pressure vessel steels. PVRC-sponsored interpretive report covers the nature of this phenomenon, its manifestation and how it may be prevented

By LEONARD D. JAFFE

SUMMARY. Temper brittleness is an embrittlement of carbon and low-alloy steels arising from exposure to temperatures above 700° F but below those at which austenite forms. It is characterized by a rise in the temperature of transition from brittle to ductile failure, with no change in strength or hardness. The severity of temper brittleness in the low-carbon pressure vessel steels commonly used in the as-rolled or normalized condition is generally not appreciable, and no failures from this source have been reported. The effect may be more severe in quenched and tempered low-alloy steels, particularly those containing manganese, chromium and, to a lesser extent, nickel.

The mechanism of temper brittleness is not known, but the conditions under which it is produced are fairly well defined. The classic example of temper brittleness is that arising in quenched and tempered alloy steels on slow cooling from the tempering temperature.

In general, the obvious method of preventing temper brittleness is to avoid the temperature conditions under which it develops. In quenched and tempered steels, this means, for example, tempering at a high temperature for a short time rather than at a lower temperature for a longer time. Water quenching from the tempering temperature is often practiced. Heating above the austenitizing temperature eliminates brittleness resulting from previous treatments. Molybdenum up to about 0.60% is also useful in greatly retarding temper embrittlement.

The values obtained from the conventional tensile test give little indication of the susceptibility of a steel to temper embrittlement. The usual method of measuring susceptibility is to use the notched-bar impact test, and to compare the transition temperature of specimens slowly cooled from a tempering temperature of about 1100° F with the transition temperature of specimens water quenched from the same temperature.

Materials Division
Pressure Vessel Research Committee
Welding Research Council

Introduction

Temper brittleness is an embrittlement of carbon and low-alloy steels arising from exposure to temperatures above about 700° F but below those at which austenite forms. In service or test, temper brittleness is characterized by a rise in the temperature of transition from brittle to ductile fracture with no change in hardness or strength.

The temperature of transition, in either unembrittled or embrittled condition, cannot be given as a definite number since it depends on the rate of loading, stress conditions including notches, prior deformation, composition, grain size, deoxidation practice, microstructure and cleanliness. For any fixed combination of these variables, however, temper brittle steel will behave in a brittle manner at higher temperatures than will unembrittled steel (Fig. 1).

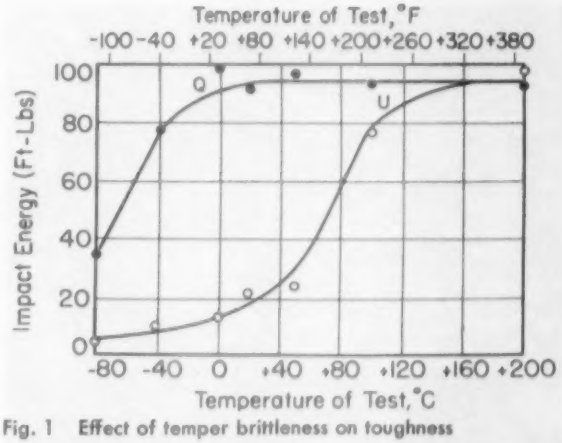


Fig. 1 Effect of temper brittleness on toughness

V-notch Charpy impact test. SAE 3135 steel, austenitized 1 hr at 1600° F, water quenched, tempered 1 hr at 1110° F, water quenched. Left curve (Q), unembrittled. Right curve (U), embrittled 500 hr at 850° F, water quenched. After Hollomon, Jaffe, McCarthy and Norton.

The service or test temperature range at which embrittlement is manifest may be below, at or above room temperature, depending on the variables mentioned. Temper brittleness in a steel may be distinguished from other types of embrittlement in three ways:

1. Fracture below the transition temperature range is predominantly intercrystalline (along grain boundaries) rather than transcrystalline (along cleavage planes within the grains), unless the temper brittleness is mild. This characteristic may be revealed by examining the fracture through a binocular microscope (Fig. 2). It is shown more definitely by nickel-plating the fracture, cutting a section through it, and examining metallographically (Fig. 3).

Leonard D. Jaffe is Chief, Materials Section, Jet Propulsion Laboratory, California Institute of Technology, Pasadena, Calif.

This report has been prepared under the sponsorship of the Materials Division of the Pressure Vessel Research Committee of the Welding Research Council.

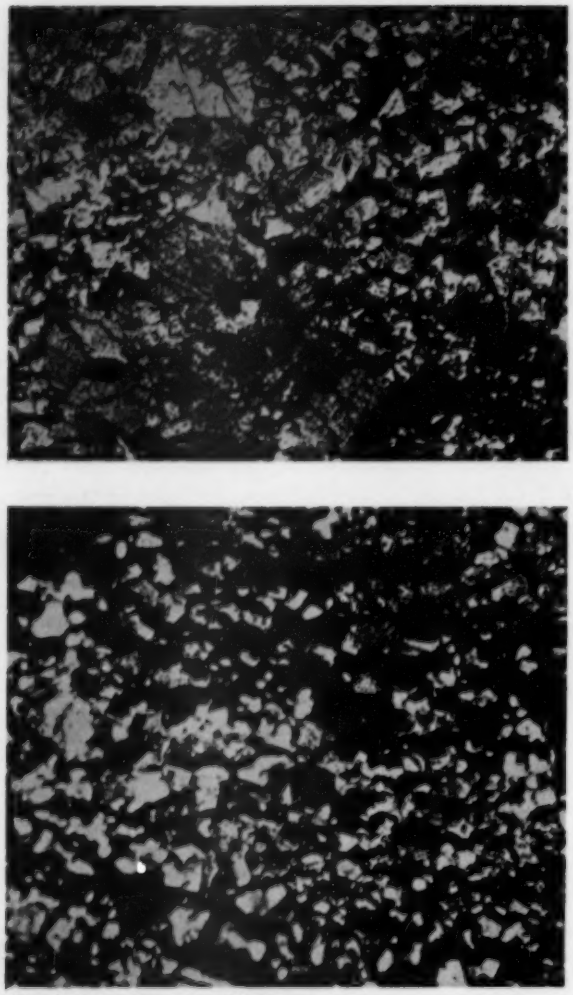


Fig. 2 Appearance of fracture

SAE 3140 steel, quenched and tempered, Charpy specimen broken at —320° F X 10. (*Top*) Unembrittled. Appearance: each facet evenly lighted, flat. Transcrystalline cleavage. (*Bottom*) Embrittled. Appearance: many of the facets show shading, curved. Predominantly intercrystalline grain boundary break. After Norton.

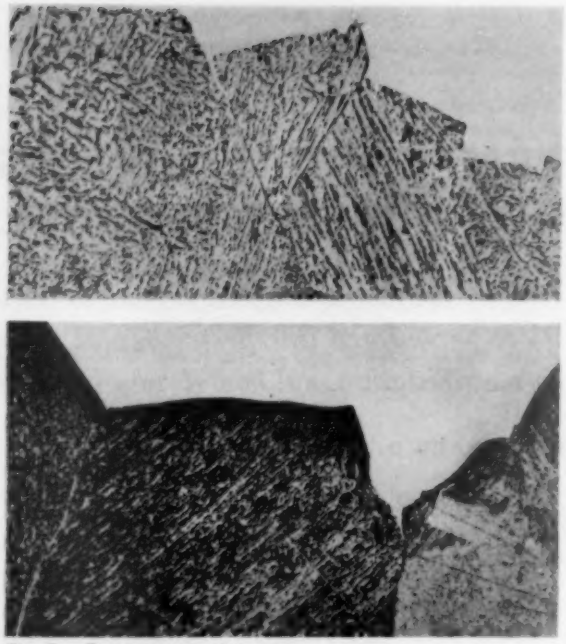


Fig. 3 Cross section of fracture

Specimens broken below transition temperature, nickel-plated, sectioned, metallographically polished and etched. (*Top*) Unembrittled steel: mostly straight segments of fracture, across grains. After Hurlich. (*Bottom*) Embrittled steel: curved segments of fracture, along grain boundaries. After Norton.

2. Temper brittle material is etched more rapidly at austenitic and ferritic grain boundaries by certain special metallographic reagents than is unembrittled material of the same composition (Fig. 4). Picric acid in water or in ether is one of these reagents, picric acid in alcohol ("picral") is *not*. This test is useful only when embrittled material of the same composition is available for comparison. The etch appears merely to dissolve away grain boundaries and does not reveal any particles in embrittled material that are not present in unembrittled (Fig. 5).

3. Hardness (strength) is unaffected by temper

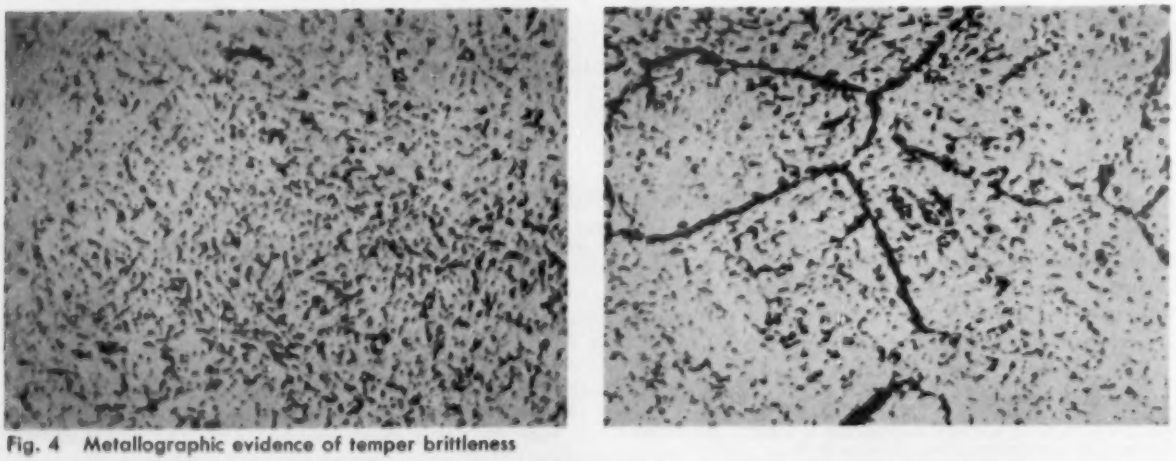


Fig. 4 Metallographic evidence of temper brittleness

Specimens etched together, with picric acid in ether plus Zephiran wetting agent. X 2000. (*Left*) Unembrittled. (*Right*) Embrittled. After Cohen, Hurlich and Jacobson.

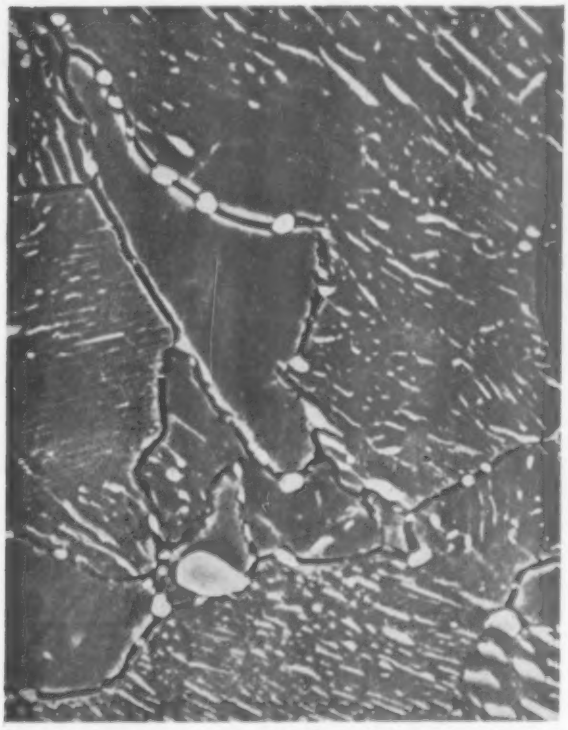


Fig. 5 Appearance of grain boundaries of embrittled steel at high magnification

Etched with aqueous picric acid. Formvar replica electron micrograph. X 7500. Note groovelike appearance of grain boundaries. After Woodfine.

brittleness. Ductility and energy absorption above the transition temperature are decreased slightly in severe cases.

The lattice parameter of the ferrite phase, as measured by X-ray diffraction, is lowered very slightly by temper brittleness. This is probably not useful for identification. The higher temperature of transition from brittle to tough fracture of temper brittle steel than of unembrittled material does not distinguish temper brittleness from other forms of brittleness. It provides, nevertheless, the only known quantitative measure of temper brittleness: the "transition-temperature-increase," usually expressed in degrees.

Thermal Cycles Leading to Temper Brittleness

The classic example of temper brittleness is that arising in quenched and tempered alloy steels on slow cooling from the tempering temperature. Temper

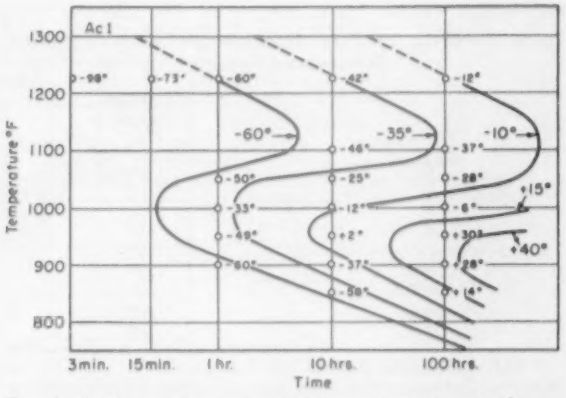


Fig. 6 Isothermal temper embrittlement of alloy steel

Heavy lines represent constant transition temperatures, given in ° F, and measured at 1/2 maximum impact energy in V-notch Charpy test, after austenitizing 1 hr at 1550° F, water quenching, tempering for 1 hr at 1225° F and embrittling for times and temperatures indicated on graph. (Prior temper was omitted for points shown at 1225° F on graph.) 0.26% C, 1.6% Mn steel. After Powers and Carlson.

brittleness also arises on holding at constant temperature.

Figure 6 illustrates how the development of temper brittleness depends upon time and temperature for a typical alloy steel. Temper brittleness may develop at all temperatures below the A_1, above which austenite forms. There are two ranges in which embrittlement is most rapid: 900–1000° F and above 1200° F. Below 800° F embrittlement is slow; near 1100° F it is also slow.

Embrittlement in the upper embrittlement range increases with time. Embrittlement in the lower range increases for at least 2500 hr in some steels; in others, the transition temperature levels out or perhaps starts to decrease in much shorter time. Accordingly, it is uncertain whether an equilibrium degree of temper brittleness is eventually attained at each temperature or whether restoration of toughness eventually takes place. The problem is complicated by the softening due to tempering during long holds, which tends to lower the transition temperature. As a practical matter, long time holding at temperature cannot be counted upon to reduce temper brittleness.

Temper brittleness develops more rapidly during cooling through a range of temperatures than on holding at any constant embrittling temperature within the range. It can also develop on heating (Table 1).

If a period of exposure to elevated temperature is interrupted by one or more quenches to room temper-

Table 1—Development of Temper Brittleness During Cooling and Heating*

	Embrittling treatment				
Cooling from temper	*Heating method*	*Temp., ° F*	*Time, hr*	*Cooling*	*Transition temp., ° F*
30°/hr to 930° F	....	930	48	30°/hr	+95
Water quench	Salt at	930	240	Water quench	+75
Water quench	Salt at	930	48	Water quench	+30
Water quench	30°/hr to	930	48	Water quench	+60

* SAE 3140 steel, V-notch Charpy test. All specimens austenitized 1 hr at 1650 °F, water quenched, and tempered 1 hr at 1245° F. *After Jaffe, Buffum, and Carr.*

Table 2—Effect of Prior Embrittlement Upon Transition Temperature Increase*

Transition temperature without slow cool, ° F	*Transition-temperature-increase, ° F, produced by slow cool from 1245 to 950° F*
V-notch Charpy test	
−110	110
−85	70
−65	80
−60	80
+40	25
+75	20

* SAE 3140 steel austenitized 1 hr at 1650° F. Water quenched and tempered 1 hr at 1245° F, then given various treatments at lower temperatures to produce transition temperatures shown in left column.

After Jaffe, Buffum, and Carr.

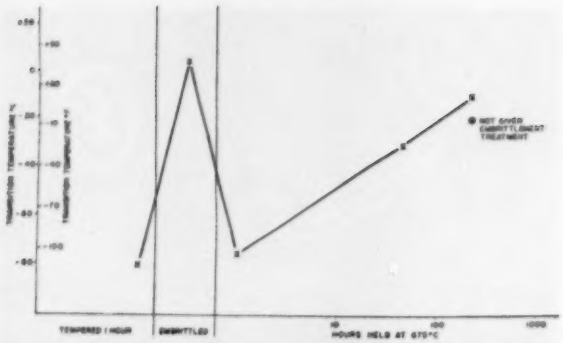


Fig. 7 Effect upon transition temperature of reheating embrittled steel near upper nose

SAE 3140 steel, quenched, tempered at 675° C (1245° F), embrittled 48 hr at 500° C (930° F), and reheated at 675° C (1245° F). After Jaffe and Buffum.

ature and rapid reheats, temper brittleness develops just as though the interruptions did not occur. If, however, the cooling or heating is slow, additional embrittlement develops.

If several different embrittling treatments are applied successively to the same material, each adds to the embrittlement. The transition-temperature-increase associated with a given treatment is less when the treatment is applied to material already severely embrittled than when applied to material not previously embrittled (Table 2).

Temper brittleness developed at low tempering temperatures (800–1100° F) can be removed by heating for a short time at 1100° F or above, and cooling rapidly. If the heating above 1100° F is prolonged, brittleness reappears (Fig. 7). Some workers in the field accordingly term embrittlement associated with the lower "nose" of the isothermal embrittlement diagram (Fig. 6) "reversible temper embrittlement" and that associated with the upper nose "irreversible temper embrittlement," and consider them to be separate phenomena, arising from separate causes. Others believe that the upper nose embrittlement should not properly be termed "temper brittleness" at all.

Temper brittleness is completely removed by reheating the embrittled steel into the austenite range and permitting it to transform to austenite.

Effect of Composition

The addition of molybdenum is commonly believed to reduce temper brittleness. It seems that molybdenum, in amounts up to 0.60%, markedly decreases the amount of embrittlement associated with the lower nose. Given sufficient time of exposure, however, brittleness develops in molybdenum-bearing steels and may become just as great as in similar steel without molybdenum (Fig. 8). The probable explanation is that molybdenum has little effect on upper nose embrittlement, and that the two types of embrittlement overlap. Thus, at very long times upper nose embrittlement may develop at 900° F or even at lower temperatures.

Molybdenum in percentages larger than 0.60–0.70% cannot be used effectively to combat temper brittleness because secondary hardening occurs. This phenomenon, the basis of high-speed steels, manifests itself as an increase in hardness, most marked after holding near 1000° F, accompanied by a drop in toughness and ductility. Secondary hardening is caused by precipitation of alloy carbides of crystal structure different from that of cementite.

The transition-temperature-increase associated with a fixed treatment in the embrittlement range becomes larger as the content of chromium (Fig. 9) or manganese is raised. Chromium above 2% also introduces

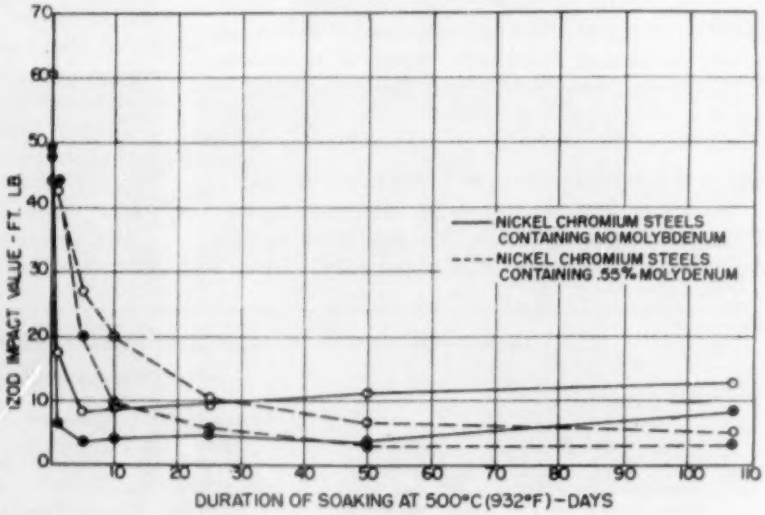


Fig. 8 Effect of molybdenum in retarding brittleness

Steels annealed at 1510° F, furnace cooled. After Lea and Arnold.

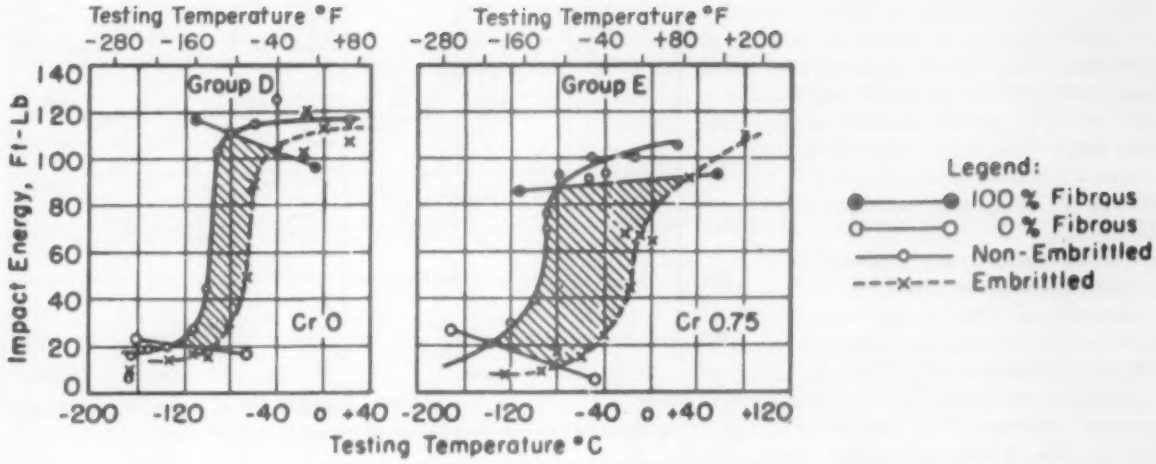
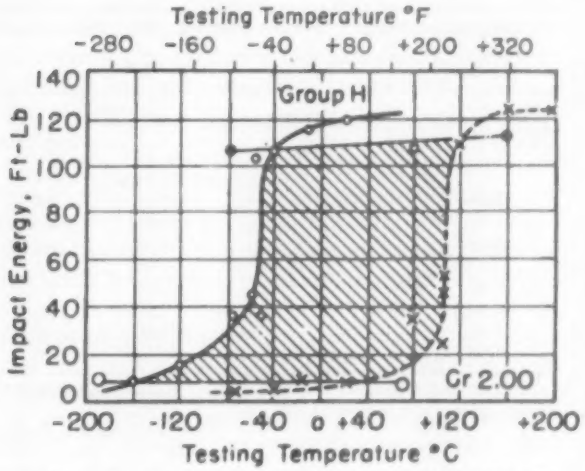


Fig. 9 Effect of chromium on temper brittleness

Austenitized 1 hr at 1625–1675° F, water quenched. Tempered 1–5 hr at 1100–1175° F. Nonembrittled specimens water quenched from temper. Embrittled specimens cooled from temper at 32° F/hr. After Tober, Thorlin and Wallace.


Testing Temperature °F
-280 -160 -40 +80 +200 +320
Group H
Cr 2.00
Impact Energy, Ft-Lb
140 120 100 80 60 40 20 0
-200 -120 -40 0 +40 +120 +200
Testing Temperature °C

secondary hardening, especially around 800–900° F. The effect of nickel is similar to the effect of manganese, but less marked.

Vanadium above 0.10% introduces secondary hardening, especially at 1000–1200° F. Because vanadium carbides dissolve slowly in austenite at low temperatures this secondary hardening is small when the prior austenitizing temperature has been lower than, say, 1700° F. If the prior austenitizing temperature has been high, the vanadium carbides will have dissolved and considerable secondary hardening will be observed.

Tungsten also is likely to lead to secondary hardening, if over 1% is present.

Iron alloys with 0.003% carbon or less do not appear susceptible to temper brittleness (Fig. 10). It may occur, however, in the presence of as little as 0.006% carbon. Increases in carbon content beyond this minimum do not seem to produce any very consistent effect upon temper brittleness.

Phosphorus greatly increases susceptibility to temper brittleness. Even 0.020% phosphorus is much worse than 0.005%, and higher phosphorus contents produce greater transition-temperature-increase (Fig. 11). Sulfur and silicon, on the other hand, seem to have no effect upon temper brittleness.

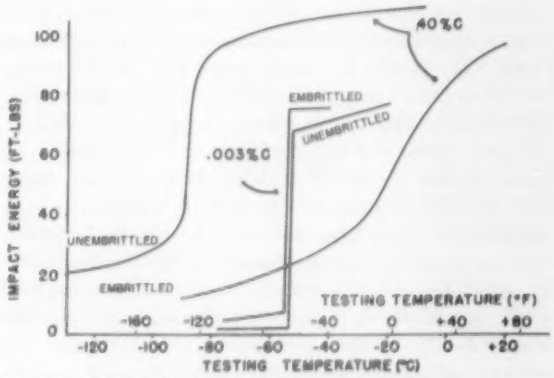


Fig. 10 Effect of carbon on temper brittleness

SAE 3100 steels. V-notch Charpy tests. Water quenched from 1650° F tempered 1 hr at 1245° F. Unembrittled specimens water quenched from temper, embrittled specimens cooled 30° F/hr from temper. After Buffum, Jaffe and Clancy.

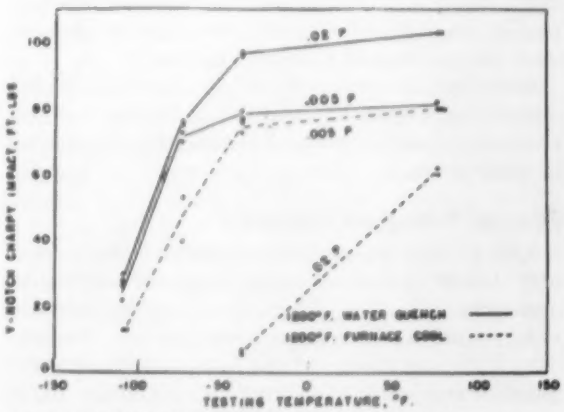


Fig. 11 Effect of phosphorus on temper brittleness

Steel—0.25% C, 1.0% Mn, 0.75% Cr—quenched from 1625° F and tempered as indicated. V-notch Charpy tests. After Herres and Elsea.

Deoxidation practice in general, and small amounts of aluminum in particular, seem to have little effect upon temper brittleness. Nitrogen and probably oxygen content also have little effect. The effect of hydrogen is not known but steels with as little as 10^{-6}% hydrogen are susceptible.

Boron raises the transition-temperature-increase associated with treatment near 1000° F, but does not affect embrittlement occurring during heat treatment over 1200° F.

Arsenic, and probably titanium, increase susceptibility somewhat. Cobalt, tin, lead and copper in small amounts have little effect on temper brittleness. Copper does lead to precipitation hardening and accompanying brittleness because of formation of a copper-rich phase, but this is not the same phenomenon as temper embrittlement.

Antimony introduces an extreme susceptibility to temper brittleness; 0.08% antimony has a very large effect. Antimony is rarely analyzed for in steels, but it appears that 0.02–0.03% usually is present. Whether this much is significant is not yet known.

Upper nose embrittlement is extremely rapid in plain carbon steels (Fig. 12). Plain carbon steels also develop significant temper brittleness in the region of the lower nose, 900–1000° F, though the transition-temperature-increase found in this region is not too great and it is not certain whether such steels show a true lower nose. High purity iron-carbon alloys are susceptible to temper brittleness, at least when the carbon content is 0.02–0.10% (Table 3).

Many of the data in the literature concerning the effects of composition upon temper brittleness are misleading. Conclusions often have been based on measurements of impact energy absorption at room temperatures only. Because changes in composition may directly or indirectly shift the transition temperature of unembrittled material, this type of measurement must be viewed with great caution.

In general, it appears that all ferritic steels are susceptible to temper brittleness, but the rate of embrittlement may be changed considerably by changes in composition.

Effect of Prior Heat Treatment

A steel is more susceptible to temper brittleness when fully hardened and tempered (tempered martensitic microstructure) than when incompletely hardened and tempered (bainitic microstructure). Susceptibility is still less when normalized, as -rolled or annealed (pearlitic and ferritic microstructure) (Table 4). Even pearlite and ferrite are significantly susceptible, however. The transition temperature of unhardened steel in the unembrittled state is so much higher than

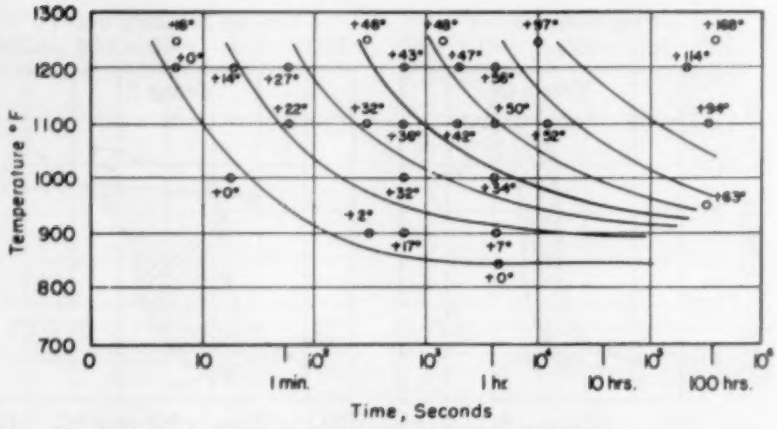


Fig. 12 Isothermal temper embrittlement of plain carbon steel

AISI 1050 steel, austenitized 1/2 hr, brine quenched, tempered 5 sec at 1200° F, water quenched. Figures beside points represent increase in transition temperature (° F) from the unembrittled condition, measured at 50% fibrous fracture in V-notch Charpy test, for embrittlement treatment at the indicated temperature for the indicated time, followed by water-quenching. After Libsch, Powers and Bhat.

Table 3—Embrittlement of High-Purity Iron-Carbon Alloys*

	V-notch Charpy specimens—Transition temperatures, ° F—		
Carbon, %	Unembrittled	Embrittled	Difference
0.025	−5	+85	90
0.045	+40	+85	45
0.07	+50	+75	25
0.11	+60	+95	35
0.17	+40	+40	0
0.24	+115	+115	0

* Quenched, tempered 1 hr at 1110° F. Embrittlement treatment: Cool from temper at 45° F/hr.

After Hultgren and Chang.

that of unembrittled tempered martensite, at the same strength level, that the transition temperature after temper embrittlement is higher for the unhardened than for the hardened material (Table 4). In fact, in the instance cited in the table, the transition temperature of the embrittled tempered martensite remains well below that of the unembrittled pearlitic steel.

Most investigations of temper brittleness have been made on quenched and tempered steels; relatively little is known of the details in unhardened steels. For example, the studies of effect of composition mentioned above were made on quenched and tempered steels. Their applicability to unhardened steel has not been checked experimentally, though there is no reason to doubt that the same rules generally apply.

Increase in austenitic grain size raises the susceptibility to temper brittleness (Fig. 13). This is in addition to the direct effect of increased grain size in raising the transition temperature of unembrittled material. A change in the temperature from which a steel is quenched, without change in grain size, seems to have little if any effect upon development of temper brittleness.

Hardness likewise appears not to affect the transi-

Table 4—Effect of Microstructure Upon Temper Brittleness*

Structure	—Transition temperature, ° F— Unembrittled	Embrittled	Difference
Tempered martensite	−65	+15	80
Tempered pearlite and ferrite	+105	+140	35

* SAE 3140 steel, Rockwell B 95–97. V-notch Charpy test. Austenitized 1 hr at 1650° F. Embrittlement treatment 48 hr at 930° F.

After Buffum and Jaffe.

tion-temperature-increase associated with temper brittleness.

Temper Brittleness in Welding

Temper brittleness develops in steel castings in exactly the same way as it does in wrought steel. Though data on temper brittleness in weldments are very limited, it would be expected that temper brittleness would develop in weldments just as it does in unwelded material subjected to the same thermal cycle.

In single-pass welding, cooling from the welding heat provides a quench for the weld metal and heat-affected zone, and the heating provides a temper for the adjacent material that is not heated hot enough to form austenite. In multipass welding, the weld metal and heat-affected zone receive both quench and temper. Preheating for welding will reduce the quench severity and, by preventing martensite formation, will reduce the transition-temperature-increase during subsequent exposure to temper embrittlement conditions. On the other hand, a stress-relieving treatment, including the accompanying heating and cooling periods, provides considerable opportunity for temper brittleness to develop.

There has been considerable concern recently over embrittlement developing in nickel-molybdenum-vanadium alloy weld metal during stress relieving. A definite hardening accompanies this embrittlement. Apparently the embrittlement is not temper brittleness but is rather due to secondary hardening associated with a vanadium content of 0.10% or more. This secondary hardening occurs after welding to a much greater extent than after ordinary heat treatment because the very high austenitizing temperatures provided by welding dissolve vanadium carbides previously present.

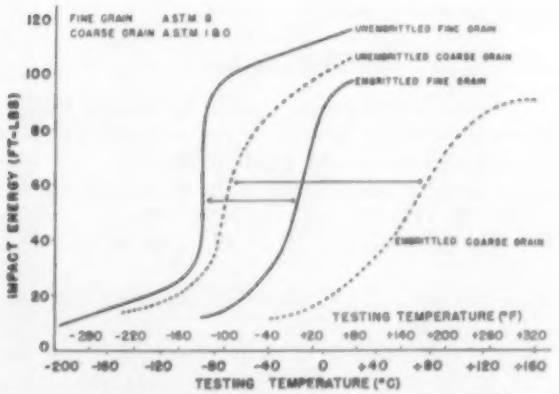


Fig. 13 Effect of austenite grain size on temper brittleness

SAE 3140 steel. Fine grain austenitized 1 hr, 1650° F, water quenched. Coarse grain austenitized 4 hr, 2300° F, water quenched. All tempered 1 hr, 1245° F, water quenched. Embrittling treatment 48 hrs, 930° F, water quenched. After Jaffe, Corr and Buffum.

Temper Brittleness in Pressure Vessels

Temper brittleness in pressure vessels might develop during stress relieving, as mentioned in the preceding section. If the service temperature is over 800° F temper brittleness can develop in service. It also may develop during forming or straightening at 800-1300° F.

Plain carbon pressure vessel steels or weld metal probably will develop almost their maximum temper brittleness if exposed for a short time to embrittling conditions. Elevated temperature service, therefore, is not likely to produce significant additional temper brittleness in plain carbon steels beyond that which may be developed by stress relieving. In chromium, manganese and nickel steels, temper brittleness might develop during stress relieving but could become much more severe during service. Molybdenum-bearing pressure vessel steels and weld metal, including chromium-molybdenum compositions, probably would become brittle only during long-time service above 800° F.

In copper-bearing steels, precipitation hardening would be expected under conditions normally conducive to temper brittleness, and it would probably be difficult to distinguish between embrittlements arising from these two causes. Steels containing vanadium are likely to develop still another type of brittleness, that associated with secondary hardening, especially in regions heated over 1700° F during welding.

Away from the welds, in pressure vessel steels, the increase in transition temperature of the base metal resulting from temper brittleness would be expected to be small or moderate, because pressure vessel steels are used in the as-rolled, normalized or stress-relieved conditions and not in the fully heat-treated condition (tempered martensite). In weld metal and weld heat-affected zones the increase may be greater because these zones are effectively quenched by rapid cooling from the welding heat, unless preheated before welding. Whether the increase in transition temperature because of temper embrittlement during subsequent stress-relieving or service is of practical importance depends not only on the amount of the increase but also on two other factors: First, what was the transition temperature without temper brittleness? If this transition temperature to begin with was sufficiently low (the material sufficiently tough), even a large embrittlement will still leave the material with adequate toughness. Second, how severe are the service conditions? If the material is likely to be subjected to shock at ambient or especially at lower temperature, particularly if the shock includes tension across notches or biaxial tension, high toughness is likely to be needed. At

the other extreme, in material loaded only when hot, temper brittleness would not matter at all. Intermediate cases, where some toughness might be required, could involve static loads at ambient or winter temperatures.

There seems to have been no reported failure of a conventional pressure vessel because of temper brittleness. It is possible, however, that failures may not have been publicly reported, or may not have been recognized as attributable to temper brittleness.

Prevention and Cure

In general, the obvious method of preventing temper brittleness is to avoid the temperature conditions under which it develops. In quenched and tempered steels, this means, for example, tempering at a high temperature for a short time rather than at a lower temperature for a longer time. Water quenching from the tempering temperature is often practiced. In pressure vessel steels, temper brittleness apparently is not produced by the heat treatments normally used, and little can be done about the heat effects of welding or of service. Forming or straightening at 800–1300°F can be avoided however. Temper embrittlement during stress relieving can be minimized by using a short time at 1150–1200° F instead of a long time at a lower temperature. Also, slow heating to the stress-relieving temperature and particularly slow cooling afterward may be minimized; it is likely that air cooling rather than furnace cooling can be employed in many cases.

Another approach to minimizing the practical effects of temper brittleness is to start with base metal and weld metal having low transition temperatures. Normalized steel is tougher than annealed, and quenched-and-tempered is tougher than either. Use of steel with sufficient hardenability to give tempered martensite under the heat-treating conditions used represents optimum practice, but is not normally applicable to pressure vessels.

Other well known means of attaining low transition temperatures include using steels fully killed with proper deoxidizing practice, and maintaining fine austenitic grain size.

Fine grain size and low phosphorus content have direct effects in reducing temper brittleness. Use of molybdenum up to 0.60% is also helpful in greatly retarding development of brittleness.

If temper brittleness has developed, it is often possible in alloy steels to remove it by preheating for a short time at 1100–1200° F and then cooling as rapidly as feasible. The temperature of reheating must be above that at which the temper brittleness developed. It is not known whether this treatment is of value in plain carbon steels. Temper brittleness in any steel may be removed by reheating into the austenite range, as in annealing or normalizing; this may sometimes be feasible for pressure vessels.

Theories of Temper Brittleness

A great many theories as to the cause of temper brittleness have been advanced. Only four seem to be seriously considered by other than their originators at the present time. The first is that of precipitation of iron carbide particles or films, probably submicroscopic in size, at austenite and ferrite grain boundaries. Removal of embrittlement by heating to a higher temperature in the ferrite range is then ascribed to increased solubility of the carbon in ferrite at higher temperatures.

The second theory is that temper brittleness develops because of the segregation to the grain boundaries of various elements dissolved in the ferrite. Such segregation would occur because it lowers the interfacial free energy associated with the boundaries, and would decrease at high temperatures because of entropy considerations.

The third theory attributes temper brittleness to growth of ferrite grains. This would increase at high temperatures and the theory is therefore pertinent only to so-called "irreversible temper brittleness" or "upper nose brittleness."

The fourth theory associates temper brittleness with growth of the pre-existing carbide particles, visible under the miscroscope. This theory is subject to the same limitations as the third.

None of these theories appears to be wholly satisfactory in explaining the observed facts. Decision among them, or establishment of a better theory, awaits further research.

Suggested Research

Research on temper brittleness that would be of most applicability to pressure vessel steels would include investigation of the rate and extent of embrittlement in unhardened steels, especially plain carbon steels. Short cycles, similar to those produced by welding in adjacent zones, should be included. Investigation of the possibility of removing embrittlement by treatments below that at which austenite forms is also needed in such steels. Determination of the cause of temper brittleness is clearly desirable. Tests of pressure vessel steels, including weld areas, for toughness before and after elevated temperature service would be worth while.

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Pertinent references on temper brittleness follow. Those of particular interest on various topics are:

General reviews: 50, 72, 92, 164, 173.

Metallographic identification: 33, 40, 85, 88–91, 99, 108–110, 117, 124, 142, 144, 170, 174.

Fracture identification: 34, 44, 87, 88, 91, 94, 110, 124, 133, 174.

Other identification: 23, 39, 47, 57, 65, 90, 91, 110, 117, 124, 138, 167.

Thermal cycles leading to temper brittleness: 6–8, 10, 31, 57, 95–97, 110, 112, 136, 141, 165, 174, 175.

Effect of prior heat treatment: 26, 27, 30, 35, 69, 70, 98, 101, 105, 136, 142, 174.

Effect of composition: 6–8, 12, 28, 52, 69, 70, 79, 105, 110, 141, 142, 144, 157, 162.

Temper brittleness in plain carbon steels: 44, 61, 70, 94, 112, 142, 161.
Secondary hardening: 11, 143, 153, 165, 169.
Temper brittleness in welding: 132, 143, 153, 169.
Theories: 2, 5, 29, 34, 62, 79, 93, 103, 125, 146, 174.

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* References marked with asterisk are apparently not concerned with temper brittleness, despite its mention in their title.

DISCUSSION BY H. S. BLUMBERG

There is a continual need for more data and information on the behavior of steels in fabrication and service and, therefore, the Interpretive Report on Temper Embrittlement of Pressure Vessel Steels dealing with a phenomenon which may result in a deterioration of materials quality will be welcomed by the Pressure Vessel Industry. Dr. Jaffe is known for his research studies and publications on temper embrittlement over a number of years. The subject is a difficult one to treat in an Interpretive Report, mainly, as the author states, because the mechanism of the behavior is not known. Dr. Jaffe has, however, made available considerable information and the treatise will be of value to metallurgists, engineers and users of materials.

The writer offers the following comments on the Report based on a number of years of metallurgical experience with the fabrication and service behavior of pressure vessels and piping in the petroleum, steam power piping and chemical industries. This discussion is intended to be constructive and for the purpose of supplementing the Report, particularly in terms of the meaning of the temper embrittlement phenomenon to the actual user of pressure vessels and piping.

1. The Report would be more significant if the author had related temper embrittlement to at least one service experience, to indicate to engineers and users of equipment that the behavior is more than merely a laboratory phenomenon. Such a dramatic example is the behavior in the pressure vessel field over twenty years ago of 5% chromium steel (containing no molybdenum) when that single alloying element was first added to plain carbon steel to combat sulfur corrosion at elevated temperatures in the petroleum refining industry. Fabricated piping given an annealing heat treatment was supplied for service with notched-bar (Charpy V) impact values at room temperature over 100 ft-lb. After several months' operation at approximately 1000° F, the piping was cooled to room temperature for shutdown to allow coke to be removed by turbine cleaning (high impact). Severe cracking of the piping occurred accompanied by brittle fracturing. Impact tests showed considerably less than 15 ft-lb absorbed energy. By using steam to raise the pipe temperature to about 200° F, the coke cleaning could be performed without failure. It was this actual service experience which was responsible (in USA) for the addition of at least ½% of molybdenum to the plain chromium steels which element was shown first to retard temper embrittlement, and later, in addition, to increase high temperature strength at certain elevated temperatures.

2. Dr. Jaffe states in Summary that the classic example of temper embrittlement is that arising in quenched and tempered alloy steels on slow cooling from the tempering temperature. Some may interpret this to mean that quenching and tempering are necessary for severe effects to take place. The 5% Cr steel case cited in (1) above, in which the steel was in the annealed condition, shows that quenching and tempering are not a necessary requirement for severe temper embrittlement.

3. Although Dr. Jaffe implies it, it is worth while emphasizing that while temper embrittlement results from certain elevated temperature treatments, the effects are manifested near atmospheric temperatures (transition temperature) so that equipment made of steels susceptible to temper embrittlement does not appear to be affected while operating at elevated temperatures.

4. The word "embrittlement" is not defined sufficiently in the Report and will, undoubtedly, cause alarm on the part of many designers, fabricators and users of materials. It would be well, therefore, to emphasize that the statements of the author regarding "embrittlement" refer wholly to the behavior of small notched-bar specimens in an impact machine in the laboratory and that there is considerable disagreement among authorities at present as to significance of laboratory data in translation to the behavior of massive parts. Therefore, the word "embrittlement" in the text should be interpreted to refer essentially to laboratory test results.

5. The author makes a number of statements which are not consistent with the long time performance of certain materials in actual service, so that if he is taken literally, a great increase in cost of finished parts would result if industry were to adopt "precautionary measures" based on Dr. Jaffe's generalities. Thus, Dr. Jaffe refers to the "embrittlement of carbon and low-alloy steels arising from exposure to temperatures above 700° F, but below those at which austenite forms." Thousands of carbon steel pressure vessels have been subjected to such temperatures in fabrication and service without a single temper embrittlement failure reported. Dr. Jaffe suspects that temper embrittlement "may also develop during forming or straightening at 800–1300° F," and concludes that "forming or straightening at 800–1300° F can be avoided." Again, it should be noted that thousands of plates and pipes of plain carbon and low-alloy steels have been treated thus in fabrication during the past twenty years without any failure in service. Of particular concern to many will be the statement that "molybdenum bearing pressure vessel steels and weld metal, including chromium molybdenum compositions, probably would become brittle only during long time service above 800° F." Many Cr-Mo steel units have been operating at these elevated temperatures for periods of well over ten years without difficulty and these steels continue to be specified for the range between 900 and 1100° F for long time service. Based on a large background of service experience with plain carbon and

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low-alloy steels currently used, there is not the slightest indication that "temper embrittlement" is occurring in degree sufficient to warrant serious concern by industry at present.

6. Dr. Jaffe has performed a service in summarizing the current fundamental knowledge and in suggesting further research. There is, however, a need for the organization of the applications phases of the subject so that designers, materials producers, fabricators and users will have a general reference guide as to the behavior of specific compositions in both laboratory testing and, where available, in service. A first approach might be in the classification of specific steels which are used in pressure vessel service based on laboratory test results as follows: (1) mildly embrittled, (2) moderately embrittled, (3) severely embrittled. This grouping could be correlated with available service behavior. It is my experience that steels which would be classified as "mildly" or "moderately" temper embrittled are behaving without any difficulty in service and that, based on experience to date, Industry need be cautious only with the use of those steels classified as "severely" embrittled in laboratory testing.

7. More knowledge of the subject is, however, still needed. Since temper embrittlement is a time-temperature phenomenon, the effects of relatively short time at elevated temperatures (related to fabrication procedures) and long time at elevated temperatures (related to operating conditions) should be evaluated for specific steels in future studies. The removal of "probe" specimens from high-temperature service installations with known and preferably long time periods followed by laboratory testing would probably add considerable to present knowledge.

AUTHOR'S REPLY

Mr. Blumberg's comments are indeed pertinent; I am in complete agreement with most of them.

As Mr. Blumberg states, considerable trouble was encountered with temper embrittlement of chromium (and nickel-chromium) steels for elevated temperature piping. Most of the British articles listed in the bibliography for the period 1932–38 deal with these difficulties.

"Embrittlement" is used in the Report in the generally accepted sense of "increase in temperature of transition from brittle to tough fracture." Though test data are on small notched-bar impact specimens, the increase in temperature of transition occurs also in massive parts, notched or unnotched, loaded rapidly or slowly. Whether this embrittlement will or will not cause failure in service depends on the service conditions and upon the transition temperature before embrittlement, as mentioned in the Report. This problem was discussed in detail by T. N. Armstrong, N. A. Kahn and H. Thielsch in a report, "Transition from Ductile to Brittle Behavior in Pressure Vessel Steels," sponsored by the Materials Division of the Pressure Vessel Research Committee, which appeared in the August 1952 *Welding Research Supplement*, as well as in other reviews sponsored by the Welding Research Council.

The fact that elevated-temperature pressure vessels almost invariably function satisfactorily despite the possibility of temper brittleness is pointed out in the Report. Mr. Blumberg's suggestions for further research are very much in line with my own.

RESEARCH NEWS

(*Continued from page 140-s*)

Ship Structure Committee Host To British Admiralty Scientist

T. S. Robertson, Chief of the Mechanical Testing Division of the British Admiralty's Naval Construction Research Establishment in Rosythe, Scotland, recently completed a five-week visit with leading research investigators in the United States who are working on the ship fracture problem. Mr. Robertson has developed a unique method of determining the notch sensitivity of ship plate which offers much promise over other methods. His trip was made possible by the Office of Naval Research upon the request of the Ship Structure Committee. The Ship Structure Committee is composed of representatives of the Bureau of Ships, U. S. Navy, the Military Sea Transportation Service, Maritime Administration, American Bureau of Shipping and the U. S. Coast Guard. This Committee is charged with conducting a research program for the purpose of improving the design, materials and methods of fabrication of ship structures.

T. S. Robertson bids farewell to Rear Admiral K. K. Cowart, USCG, Engineer-in-Chief of the Coast Guard and Chairman of the Ship Structure Committee. Included in the group are, from left to right, F. C. Bailey, National Academy of Sciences; Capt. C. J. Palmer, USN; Mr. Robertson; Admiral Cowart; E. M. MacCutcheon, Jr., Bureau of Ships; and Commander D. B. Henderson, U. S. Coast Guard,

FATIGUE OF SPOT WELDS

Review of reports relating to the fatigue resistance of untreated and hydrostatically treated spot welds and its chronological development during the last 10 years

BY GEORGES WELTER

1—Problem of Low Fatigue Resistance of Spot Welds and Early Research Work

The problem of improving the fatigue resistance of spot welds was raised about 10 years ago by the Fatigue of Spot Welds Committee of the Welding Research Council.

If tested statically under tension, which produces prevailing shear loads, the spot welds show a satisfactory resistance. However a different picture is obtained when they are placed under repeated loads of a dynamic character, such as found in all known structures as motor cars, locomotives, railway and street-car carriages, airplanes, busses, ships, etc. Under fatigue loads, the strength of this type of welds is greatly affected, and the fatigue limit is surprisingly low. Practical experience has shown that in fatigue, riveted joints are more durable than spot-welded joints.

Tests show that under fatigue loads only about 16% of the average static shear strength of spot-welded aluminum alloys can be attained in single spots, and double spots seem to support even less than 10% of the average shear strength. Hess[1] in 1946 found, for a fatigue life value of 5,000,000 cycles the range of stress at point of failure is about 13 to 15% of the average static shear strength (the type of loading applied to these specimens consisted of a pulsating tensile stress on the sheet, superimposed on a steady tensile stress amounting to 10% of the average static shear strength.)

In a paper on fatigue of aluminum alloys Forrest[3] reports, in 1946, on fatigue tests made on single-spot welds of Alclad sheets DTD 546A of 0.036 in. thickness and carried out under fluctuating tension. The strength of these spots at 20,000,000 cycles was about 12% of the ultimate shear strength. After preloading these welds with a static load, equal to two-thirds of the average static stress at point of failure, the fatigue strength is about 18% of the ultimate static shear strength.

Furthermore, in 1947, Hartmann[2] reports on completely reversed fatigue tests of Alclad 24ST alloy spots of 0.064-in. thick sheets, which at 10,000,000 cycles support an average load of only 50 to 60 lb per spot; this fatigue limit represents less than 10% of the static shear strength. Similar fatigue test results have been recorded by Templin in 1947. The endurance limit for aluminum alloy 38 spot-welded sheets of 0.064 in. thickness, subjected to complete reversal of stress, has been found to range from 9 to 13% of the static shear strength. He states, furthermore, that "These ratios are not as great as those normally obtained for riveted lap joints in aircraft materials."

2—Hydrostatic Treatment of Spot Welds

In view of this surprisingly low resistance of spot-welded joints submitted to repeated loads, the trend was to find some way to make spot-welded structures much more resistant against vibration and repeated loads in order to improve these spots in such a way that they may be able to resist higher repeated loads and to become as durable as riveted joints.

In a preliminary test series carried out at École Polytechnique[4] in 1946–47, it was found that the number of cycles absorbed by specially treated specimens was 10 to 12 times greater than that one of untreated specimens tested under the same fatigue load.

The treatment for the improvement of electrically welded joints consists principally in a mechanical process by which extremely high pressures are applied on both sides of the weld after the welding operation is accomplished and the weld has cooled down to room temperature.

These extremely high pressures applied over the whole area of the weld are of the hydrostatic type and they exceed the elastic limit of the weld appreciably. By this mechanical high-pressure treatment the weld in the fused area of the spot is changed from a cast and annealed crystal structure into a wrought and cold-worked material of much higher mechanical properties, especially those concerning its fatigue resistance. Also the numerous imperfections in all cast metals and alloys, as crackes, fissures, porosity, transcrystallization, pinholes, segregation, harmful residual stresses, etc., are seriously reduced in the weld by this process. The mechanically treated homogeneous crystal structure in the spot area is given greatly improved yield strength and a fatigue resistance very much higher than that of untreated welds.

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Presented at Fatigue of Spot Welding Sub-Committee meeting Nov. 3rd., held in Chicago, Nov. 1–5, 1954.

The mechanical treatment for the improvement of welds consisted in the application of high hydrostatic pressure over the area of the spot weld. Starting with compression loads of about 86,000 psi., they were gradually increased to 230,000 psi. This fairly high compression stress does not, however, seem to be the optimum pressure possible of attainment with regard to the achievement of the highest fatigue resistance.

3—Improvement of the Fatigue Resistance of Spot Welds

The exceptionally high hydrostatic pressure, as applied to the welds during this investigation, seems not to improve appreciably their static resistance in tension. However, from the numerous fatigue tests carried out with specimens in the as-received condition as well as after having undergone hydrostatic pressure, it can be stated that an appreciable improvement of the fatigue properties has been recorded. These fatigue tests as carried out on two different types of pulsators are in fairly good accord, indicating that the fatigue load of mechanically treated specimens can be increased materially without sustaining failure of any welded Alclad 24ST aluminum alloy samples. The untreated samples, however, for which the lower fatigue limit could not yet be established, failed under a fatigue load of only 300 lb after a few hundred thousand cycles. This endurance load should be, according to previous results, of the order of about 10 to 16% of the ultimate static load, which may in this case represent repeated loads of about 100 to 200 lb. On the other hand it was possible to show that hydrostatically compressed welds absorbed much higher fatigue loads. At the conclusion of these tests, at the end of 1947, a load of 0 to 650 lb maximum tension was reached, and sound mechanically improved specimens did not fail under repeated loads of several million cycles. The limit of cycles for these information tests was between 4,000,000 and 5,000,000 up to 10,000,000 and in one special case 20,000,000 times this load has been repeated without any visible change in the welded spot. From the above results, temporarily the following conclusions has been drawn: an improvement of about 250 to 300% of treated spots compared to spots in the as-received condition seems possible.

These tests should, however, be extended to still higher fatigue loads as well as to a higher number of cycles at least in a few cases, to determine the final endurance limit of these welds at over 10,000,000 to several hundred million cycles.

Two years later, in 1949, the results of two investigations about spot welds were published. Otto Graf reported some results with single- and multiple-spot-welded low-carbon steel joints tested under the fatigue stresses. No attempt was made to improve the fatigue resistance of these joints by treating the specimens after welding. In the work by Welter,[5] entitled "Fatigue Tests of Spot-Welded Steel Sheets," the following conclusions were reached.

Under high compression loads of about 15,000 to 20,000 lb, applied with a ram of 1/4 in. diam, representing a maximum of about 400,000 psi, an improvement of the number of fatigue cycles of 2 to 3 times has been recorded on two different pulsators under pull-pull loads of 500 to 760 lb, applied to welded mild steel specimens.

Stainless steel specimens 18–8 of 0.03 in. thickness, 1/4 hard, tested on two different fatigue machines, supported a higher repeated load up to failure than mild steel specimens; under 650 to 770 lb the treated specimens were generally more resistant and, according to the mechanical treatment, cycles of 3 to 10 times those of untreated specimens could be recorded.

By an artificial spot, joining two mild steel sheets of 0.03 in. thickness over an area corresponding to that of an ordinary spot without using any heat, it could be shown that the so-called heat-affected zone around the spot does not weaken its fatigue resistance. Failure by cracking has taken place after about the same number of cycles as resistance-welded sheets.

The low fatigue resistance of spot-welded mild steel sheets is very likely due to a combination of tension and bending stresses rising to a fairly high amount exactly in the zone adjacent to the spot on the tension side of the applied load.

4—Investigation of Stress Distribution Around the Spots

In 1950, two papers about the stresses distributed around spot welds were published, one by the Franklin Institute of Philadelphia,[6] and the other one by the Research and Strength of Materials Laboratory in Montreal.[7]

The finding of these two reports can be summarized as follows: These investigations so far lead to a conclusion that single laps of spot-welded joints are inherently weak due to the unavoidable bending of the plates; that in important structures double laps (one plate sandwiched between two plates) will reduce considerably the high stress concentrations set up at the weld spot; that the joining of two plates by one row of resistance spot welds is inherently unsound when used for important structures carrying greatly variable loads or reversal of stress.

Spot-welded mild steel specimens put under tension loads up to 600 lb usually indicate, at the outside surface around the spot, compression stresses; slight tension stresses were measured only at a small distance above the spot in the axis of the joint.

Also with spot-welded or riveted Alclad 24ST specimens tested under identical conditions, the results showed mainly compression stresses on the surface around the spot.

Based on a greater number of results, a certain stress pattern taking place around the spot, under various loads up to 600 lb could be established.

Measurements made at the interface of the joint, instead of its out side surface, show that, due to bending of the sheet, exclusively tension stresses exist above

the spot, having fairly pronounced values; permanent deformation starts at loads of the order of 200 lb in tension.

Test results concerning the development of fatigue cracks under cyclic tension load show that these cracks start at the interface of the weld, due to repeated bending of the sheet in a zone close to the spot. These cracks develop a long time before they become visible at the outside surface of the sheet.

Fatigue results obtained with mild steel specimens of 0.04 in. thickness, obtained on three different pulsators, show that hydrostatically treated spots have a 25 to 50% higher fatigue limit than untreated joints. Similar results have been obtained with spot-welded, 0.03-in. thick stainless steel sheets.

Hydrostatic compression of the spot is even more effective on 24ST Alclad aluminum alloy sheets; an improvement of 80 to 100% of their endurance limit is possible.

5—Results Obtained with Multiple Welded Joints

Another work about the fatigue strength of spot-welded light alloy joints has been published in 1951 by Hiroshi Kihara; in this paper, the results of untreated multiple spot-welded joints are reported and the effects of variations in load, plate overlap, pitch of spots, distance between rows and span on the fatigue strength of spot-welded light alloy joints have been studied.

Furthermore, in 1951–52, two more investigations have been carried out at École Polytechnique. The results have been published at the beginning of 1954 and in the first paper deals with Alclad 24ST multiple spot welds and in the second one with Stress Distribution and fatigue resistance tests of spot welds in Cor-Ten and mild steel.[8, 9]

The hydrostatic treatment of multiple spot-welded joints has, as for single spot welds, a marked influence on their fatigue limit. By a hydrostatic compression of muliple joints of about 150,000 psi in the area of the spot, an improvement of about 80 to 100% of their endurance limit has been obtained. Still better results should be possible by a careful study of the optimum hydrostatic pressure application and the best possible shape of the die in form of a very resistant steel piston to produce the local compression of the individual spot applied after the welding process is completed.

After hydrostatic treatment, these joints have a fatigue limit which is an acceptable relation to the ultimate load under fatigue, approaching that of other structural materials.

Based on some theoretical considerations about the stress distribution around the spots, the following conclusions have been reached:

Under hydrostatic compression, practically no appreciable changes in the spots are visible and due to the only possible very small internal flow and rearrangement of the crystals in the material under these extremely high pressures, the fatigue limit is greatly improved. It seems as if the spot and its adjacent zones become denser by eliminating small voids and cracks in the cast and heat-affected material so that by this very limited internal flow of the material, the mechanical properties regarding the endurance limit, as this could be shown for 24ST aluminum alloy sheets, are fundamentally changed.

Much higher fatigue results have been obtained with hydrostatically treated Cor-Ten specimens. After a compression of the spot under a load of 15,000 lb, the fatigue limit at ten million cycles has been brought up to 1200 lb compared to the untreated specimens showing a limit of only 400 lb. The improvement by hydrostatic compression is in this case 200%.

For multiple-spot welds, similar improvements have been found which decrease, however, with the number of spots. For two spots in vertical arrangement, there is an improvement of about 150% while for a five-spot sample, there seems to be only a 50% improvement of the load at the endurance limit.

The ten million cycles fatigue limit of Cor-Ten joints with 1 to 5 spots in the as-received condition is of the order of about 12 to 18% of the ultimate shearing load. After hydrostatic treatment of the joints an improvement of the fatigue limit of about 50 to 200% has been found. On a load-per-spot basis, the fatigue limit decreases with the number of spots per joint; this load for a one-spot specimen is about twice that of 3-, 4- and 5- spot specimens.

6—Stainless Steel Multiple Spot Welds with and Without Hydrostatic Treatment

In a more recent comprehensive laboratory work carried out in 1951–52 and reported upon by V. N. Krivobok[10] in the Oct. 1954 *Welding Research Supplement*, multiple stainless steel 18–8 spot welds with and without hydrostatic treatment of the specimen have been tested under static and cycle loads. The objective of this investigation was to determine the fatigue resistance of various patterns of spot-welded joint to apply the hydrostatic compression treatment to these spot-welded specimens in order to improve their fatigue resistance and to study the stress distribution around the spots under static tension testing of the specimen.

The results obtained in this work have been summarized as follows:

(*a*) For stainless steel sheet as-spotwelded, as the number of spots per sample is increased, the load (per spot), to cause localized failure through fatigue is decreased.

Also, the course of the fatigue curve becomes steeper as the number of spots is increased and as the thickness of the sheet is increased.

(*b*) The location of the fatigue crack in the as-spot-welded tests was never in the weld button but in the heat-affected zone of the parent (sheet) metal. There is no correlation between fatigue and shear strength. A comparison of tensile (shear) and fatigue strength is not warranted since in one case the values

are representative of cast metal (weld nugget) and in the other case the values represent the resistance of heat-affected (annealed) sheet material to failure through fatigue. It has been confirmed, however, that the strength of the spot-welded joints tested in fatigue when compared to the strength of similar joints tested in straight tension is very low.

(*c*) Hydrostatic prestressing shows a definite improvement in fatigue properties over untreated specimens and this improvement exceeds that observed on other materials tested under similar conditions. This is the case regardless of the number of spot welds prestressed in one treatment.

(*d*) Simple compression of the spot welds produced a considerable improvement in fatigue over uncompressed samples of the same material.

(*e*) Similar improvement in fatigue characteristics could also be gained by prestressing by tension or peening.

This concludes the survey of the results about the fatigue resistance of untreated and hydrostatically treated spot welds obtained during the last 10-year period of research.

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2. Hartmann, E. C., and Stickley, G. W., "Summary of Results of Tests Made by Aluminum Research Laboratories of Spot-Welded Joints and Structural Elements," *Ibid.*, **26** (4), Research Suppl., 233-s to 251-s (1947).
3. Forrest, G., "Fatigue of Aluminum Alloys," *Jnl. Inst. Metals*, 1-17 (January 1946).
4. Welter, Georges, "Fatigue Tests of Spot Welds: Improvement of Their Endurance Limit by Hydrostatic Pressure," THE WELDING JOURNAL, **27** (6), Research Suppl., 285-s to 298-s (1948).
5. Welter, Georges, "Fatigue Tests of Spot-Welded Steel Sheets," *Ibid.*, **28** (9), Research Suppl., 414-s to 438-s (1949).
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7. Welter, Georges, "Stresses Around a Spot Weld Under Static and Cyclic Loads," *Ibid.*, **29** (11), Research Suppl., 565-s to 576-s (1950).
8. Welter, Georges, and Choquet, André, "Stress Distribution and Fatigue Resistance of Alclad 24ST Multiple Spot Welds," *Ibid.*, **33** (2), Research Suppl., 91-s to 98-s (1954).
9. Welter, Georges, and Choquet, André, "Fatigue Tests of Spot Welds in Cor-Ten and Mild Steel," **33** (3), Research Suppl., 134-s to 140-s (1954).
10. Welter, G. Krivobok, V. N., and Choquet, A., "Effects of Prestressing on Fatigue Strength of Spot-Welded Stainless Steels," *Ibid.*, **33** (10), Research Suppl., 509-s to 523-s (1954).

(*Continued from page 131-s*)

AUTHORS' REPLY

Mr. Audige's correlations of U-notch and Charpy V-notch test results are of considerable interest. It appears that the U-notch specimen is equivalent to the Charpy keyhole-notch specimen. The same type of energy transition curve is developed, featuring a very slight fall over a wide temperature range and then a "scatterband" range, marking a rapid decrease in energy absorption over a fairly limited temperature range. The relationship of the scatterband range in the keyhole or U-notch test to the Charpy V-test 10 ft-lb transition is the item in question. The importance of this relationship lies in the fact that the Charpy V 10 ft-lb transition temperature has been established to be a significant criterion of fracture initiation in large welded structures. If it is desired to use the keyhole or U-notch tests, it is possible to establish the approximate Charpy V 10 ft-lb temperature by a ΔT correction factor.

The authors have demonstrated that the ΔT factor varies with the type of steel. To answer Mr. Audige's first question, we have agreed with other investigators that:

$$T_{KM.S.} = T_{V10} - 20° \text{ to } 30° \text{ F}$$

for semikilled and rimmed steels such as ABS-A. The keyhole transition for these steels at 15 ft-lb. (T_{K15}) is equivalent to the keyhole mid-span transition ($T_{KM\cdot S.}$); hence the formula may also be expressed as:

$$T_{K15} = T_{V10} - 20° \text{ to } 30° \text{ F}$$

As Mr. Audige points out, there is no single, straight line relationship covering all steels. Our criticism was directed at the rapidly developing practice of applying the $-20°$ F rule to all steels, which led to considerable error not only for the estimation of the approximate Charpy V 10 ft-lb temperature but, more importantly, for the estimation of the danger temperature for possible fracture initiation. We find that a $-50°$ F correction is common to the fully killed steels.

It is significant that Mr. Audige likewise finds a different relationship between U mid-span (or U 11 ft-lb) and Charpy V 10 ft-lb depending on deoxidation practice of French steels. As indicated by his table, the rimmed steels require a $-36°$ F correction while the killed steels require a $-54°$ F correction.

In conclusion, we do not wish to be misunderstood with respect to the keyhole (or the U-notch equivalent) vs. Charpy V question. We believe that the Charpy V test is clearly superior for purposes of correlation with service performance.

ENERGY CRITERIA OF FRACTURE

Modifications of the Griffith theory are presented to cover the case for a rapidly running crack and for starting up a stationary crack

BY E. OROWAN

SUMMARY. It is shown that, for fully brittle materials, the Griffith equation represents a necessary and sufficient condition of tensile fracture. The present writer's modification of the Griffith equation is a necessary and sufficient condition of fast crack propagation in low-carbon steels; in order to initiate and accelerate a cleavage crack under static load, however, an additional condition (e.g., one demanding a certain initial plastic deformation) must be fulfilled.

The Griffith energy principle from which the Griffith equation is derived cannot be applied to ductile fracture, except when the plastic deformation is confined to a thin layer of material at the surface of fracture.

The Griffith Energy Principle

In the course of the last few years, it has become clear that the Griffith equation for the tensile strength of a brittle solid cannot be applied in its original form to brittle fracture in normally ductile steels. X-ray back refection photographs show[1] that a thin layer at the surface of apparently quite brittle fractures of low-carbon steels contains significant plastic distortion; the plastic work p in this layer amounts to roughly 2×10^6 ergs/cm^2 if the fracture has occurred not too far below room temperature. Compared with this value, the surface energy (a few times 10^3 ergs/cm^2) is negligible; consequently, if an expression of the Griffith type can be used at all in this case, the surface energy (representing the work for creating unit area of the surface of fracture) has to be replaced by the plastic surface work p. Thus, the crack propagation condition would be[2]

$$\sigma \approx \sqrt{Ep/c} \qquad (1)$$

The presence of considerable plastic distortion at the surface of fracture raises the question under what conditions the Griffith principle of virtual work can be applied to fractures accompanied by plastic deformation. This principle can be stated in the following manner: Let dW be the free energy required for increasing the length* of a crack from c to $c + dc$, and $-dU$ the elastic energy released simultaneously in the specimen if this is held between rigidly fixed grips so that the external forces cannot do work. The critical length of the crack above which it can propagate spontaneously is then determined by the condition

$$dW = -dU \qquad (2)$$

It is easily seen that the assumption of rigidly fixed grips is not essential; the same result is obtained if the crack propagation is assumed to occur under constant load. Let $M(c)$ be the elastic compliance, i.e., the reciprocal spring constant, of a specimen containing a crack of length c; thus,

$$x = MF \qquad (3)$$

where F is the tensile force acting upon the specimen and x its elastic elongation. The elastic energy of a specimen containing a crack of fixed length c is

$$U = \int_{x=0}^{x=MF} F \cdot dx = \frac{M(c)F^2}{2}; \quad (4)$$

on the other hand, the differential dU of the elastic energy when both F and c change simultaneously is

$$dU = (F^2/2)dM + MF \cdot dF \quad (5)$$

$dM = (dM/dc) \cdot dc$ being the increment of the elastic compliance due to the increase of the crack length.

If the crack length increases while the specimen is held between rigidly fixed grips ($x = MF$ = const.),

$$dx = MdF + FdM = 0 \qquad (6)$$

substitution of $MdF = -FdM$ in eq 5 gives

$$(dU)_x = -(F^2dM/2) \qquad (7)$$

On the other hand, if the crack propagates while the load is kept constant ($dF = 0$), eq 5 gives

$$(dU)_F = F^2dM/2 \qquad (8)$$

At the same time, the force F does the work

$$dL = F \cdot dx = F^2dM \qquad (9)$$

since, at constant F, $dx = FdM$.

Eqs 8 and 9 show that, if the crack propagates at constant load, half of the external work is stored as additional elastic energy of the specimen, and the other half is available for increasing the free energy of the crack walls. If the length of the crack exceeds the critical value at which eq 2 is just satisfied, the work of the applied force is more than sufficient to provide the increment of its free energy; the balance creates kinetic energy and accelerates the rate of crack propagation.

If, on the other hand, the crack propagates between fixed grips, the elastic energy of the specimen decreases according to eq 7, and its decrement is available for increasing the free energy of the crack and the kinetic energy. Comparison of eqs 7, 8, and 9 shows that the energy available for crack propagation at fixed load is the same as at fixed grips; in the former case, $-dU$ in eq 2 has to be replaced by $dL - (dU)_F$ which is numerically equal to $-(dU)_x$ for a given increment dc of the crack length.

In the present paper, two questions will be treated that have been widely discussed in connection with the brittle fracture of structural and ship steel, and on which a wide divergence of opinions has arisen. They are:

(*a*) Does the Griffith equation

$$\sigma \approx \sqrt{E\alpha/c} \qquad (\alpha = \text{surface energy}) \quad (10)$$

represent a necessary and sufficient condition of completely brittle fracture? And is the present writer's eq 1 a necessary condition of brittle fracture in low carbon steels?

(*b*) Under what conditions can the Griffith principle, eq 2, be applied to fractures involving plastic deformation?

The Griffith Equation as a Necessary and Sufficient Condition of Completely Brittle Fracture

The Griffith equation (10) expresses the condition that the elastic energy released (or the work of the applied forces) during crack propagation can

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* As in the original work of Griffith, only two-dimensional cases (cracks in plate specimens) will be considered here for simplicity. The general results can be easily extended to three-dimensional cases.

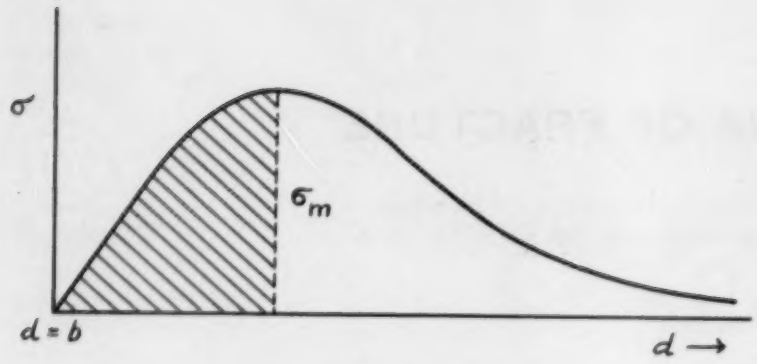

Figure 1

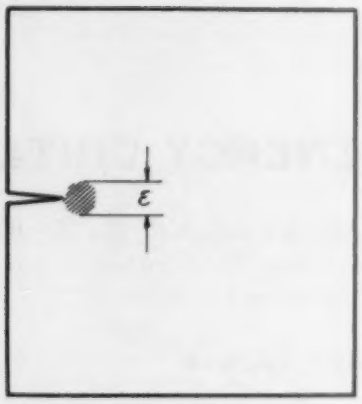

Figure 2

provide the additional surface energy of the increasing crack wall area. It is, therefore, a necessary condition of crack propagation in a completely brittle specimen under tension. If it is not satisfied, propagation of the crack with the accompanying increase of its (free) surface energy would violate the first or the second law of thermodynamics. In particular, thermal fluctuations (disruption of atomic bonds at the tip of the crack by thermal activation) cannot propagate the crack if the Griffith equation is not satisfied, because any such process would result in the creation of free energy from thermal energy without heat flowing from one reservoir to another of a lower temperature. Of course, thermal fluctuations of free energy do occur; however, they cannot lead to any significant crack propagation because the greatest energy fluctuation that may arise with any probability amounts to a few electron volts which is sufficient only for the disruption of a few individual atomic bonds at the tip of the crack. After the fluctuation has passed away, the disrupted bonds join up again.

From the fact that the Griffith equation is a necessary condition of completely brittle fracture, it does not follow that it is also a sufficient condition. However, it can be proved that once the condition is satisfied, crack propagation is not merely possible but is bound to follow. This can be shown by proving that, if the applied stress has the value given by the Griffith equation, the stress concentration at the tip of the crack reaches the value of the molecular cohesion (theoretical strength) at which fracture is bound to take place.

The molecular cohesion of a brittle material can be estimated in the following well known way. When a rod of unit cross-sectional area breaks with a smooth surface of fracture perpendicular to the axis of the rod, two new surfaces of unit area are created; the work required for this is 2α (α = surface energy). This work is done against the intermolecular attractive forces as the two fragments are pulled apart. Figure 1 shows the variation of the molecular forces between the two fragments, per unit of cross-sectional area, as a function of the distance d between the adjacent layers of molecules in the two fragments between which the fracture occurs. The force is zero when $d = b$ = the molecular spacing in the absence of stress; it rises to a maximum σ_m which is the molecular cohesion, and then falls to zero with increasing separation of the fragments. The area below the curve is the work of fracture per unit of the cross-sectional area; i.e., it is equal to 2α. At the maximum of the curve in Fig. 1, the amount of energy represented by the shaded area below the curve must be present between all neighboring pairs of molecular (or atomic) planes perpendicular to the tension; it is identical with the elastic energy stored in the material between two adjacent atomic planes. If, for an order - of - magnitude estimate, Hooke's law is assumed to be valid up to the theoretical maximum σ_m of the stress, the density of elastic energy between two atomic planes of unit area, spaced at b, is $b \cdot \sigma_m^2/2E$. If it is assumed that the shaded area is about one-half of the total area below the curve and therefore approximately equal to α, the relationship

$$b \cdot \frac{\sigma_m^2}{2E} = \alpha \qquad (11)$$

gives the order of magnitude of the molecular strength as

$$\sigma_m \cong \sqrt{2E\alpha/b} \qquad (12)$$

The next question is: what is the value of the applied tensile stress at which the critical value σ_m is reached at the tip of the crack? The stress concentration factor of a surface crack of depth c and (relatively small) root radius ρ is[3]

$$q = 2\sqrt{c/\rho}\ ; \qquad (13)$$

this relationship shows that the maximum stress would be infinitely high for any finite value of σ and c in an elastic continuum containing a perfectly sharp crack, and therefore the tensile strength would be zero. The reason why brittle solids have a finite strength lies in the atomic structure of matter. Figure 1 shows that Hooke's law breaks down when the increment of the atomic spacing becomes comparable in magnitude with the atomic spacing itself: near the tip of the crack the stress versus strain curve levels out, and the situation can be regarded roughly as if a certain region at the tip, comparable in linear dimensions with the interatomic spacing, would be under the constant stress σ_m instead of obeying Hooke's law.

This case of the law of elasticity ceasing to be valid in a region at the tip of the crack has been treated by L. Föppl[4] and, in particular, by Neuber.[5] Neuber proved the following theorem: let there be a region of linear dimensions ϵ at the tip of the crack (Fig. 2), so that the specimen is Hookean elastic outside this region, whereas the stress in the shaded region in Fig. 2 is approximately constant, having the value existing at its boundary. The ratio of the stress in the constant-stress region to the tensile stress applied to the specimen is then approximately equal to the stress concentration factor of a crack of the same length and of root radius $\epsilon/2$ in a purely Hookean elastic material (The quantity ϵ is assumed to be small compared with the length c of the crack which itself must be small compared with the dimensions of the specimen.)

In the present case, the diameter of the region in which Hooke's law breaks down and the stress levels out is obviously of the order of magnitude of the interatomic spacing b; if it is assumed to be approximately $2b$, Neuber's theorem indicates that the effective stress concentration factor is that of a crack of tip radius b in a purely Hookean specimen. According to eq 13, this is

$$q \approx 2\sqrt{c/b} \qquad (13a)$$

Thus, the value σ of the applied tensile stress at which the molecular strength σ_m is reached at the tip of the crack is given by

$$\sigma_m = \sigma \cdot 2\sqrt{c/b}\,. \qquad (14)$$

If σ_m is replaced from eq. 12, the tensile strength σ is obtained as

$$\sigma \approx \sqrt{E\alpha/2c} \qquad (15)$$

which, within the accuracy of the estimate, is identical with the Griffith eq 10.

This derivation of the Griffith equation directly from the stress concentration factor of the crack shows that, when the applied tensile stress has the value given by the equation, the stress at the tip of the crack reaches the highest value that can be withstood by the interatomic forces in the material; any further straining is bound to produce crack propagation. In other words, the Griffith equation represents not only a necessary but also a sufficient condition of fracture in a completely brittle specimen.

Can the Griffith Principle Be Applied to Ductile Fracture?

In recent years the view has been expressed that the Griffith energy principle eq 2 may be applied to all types of fracture, not only to essentially brittle ones. In what follows, it should be pointed out that this is not so: the principle can only be applied if plastic deformation is either absent or confined to a thin layer at the crack walls so that the bulk of the specimen is still elastic.

Figure 3 indicates the manner of crack propagation in a purely elastic material: owing to elastic strain release around the crack, its walls are pulled apart, and its length increases. Figure 4, on the other hand, shows one of the simplest types of ductile fracture,[6] such as is observed in aluminum single crystals or (polycrystalline) plates of ductile metals in tension. The crack (which in this case is a channel of square cross section perpendicular to the plane of Fig. 4) is propagated by slip in the planes $AB + CD$ and $EB + CF$ as indicated by arrows; in the course of this process the cross section of the crack increases until fracture is complete.

The fundamental difference between the propagation of the brittle crack shown in Fig. 3 and the ductile mechanism of Fig. 4 is that the former relies essentially on the elasticity of the material, while the latter could work in the same way even if the elastic moduli were infinitely high. The Griffith eq 10 shows directly that the tensile strength of a brittle material would rise to infinity with an infinite increase of the value of Young's modulus: in such a material, the crack could not open up because there would be no elastic strains to release. On the other hand, the slip mechanism shown in Fig. 4 is quite independent of the elastic moduli.

It is immediately obvious that the force required for propagating the crack in Fig. 4 cannot be derived from the Griffith principle eq (2). Its value is simply

$$F = Y \cdot A \qquad (16)$$

where Y is the yield stress of the material in tension and A the projection of the areas AB plus CD on the plane perpendicular to the direction of the tension. If F satisfies eq 16, the plastic deformation that opens up the crack can progress, and the crack propagates. The elastic moduli do not appear in eq 16; they could be infinitely high without any consequence to the propagation of the crack. On the other hand, infinitely high elastic moduli would make the right-hand side of eq 2 vanish: the tensile strength obtained by any application of the Griffith energy principle would rise to infinity with the elastic moduli.

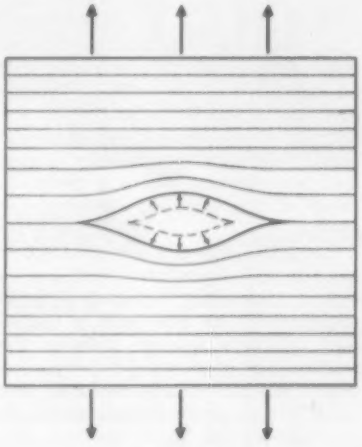

Figure 3

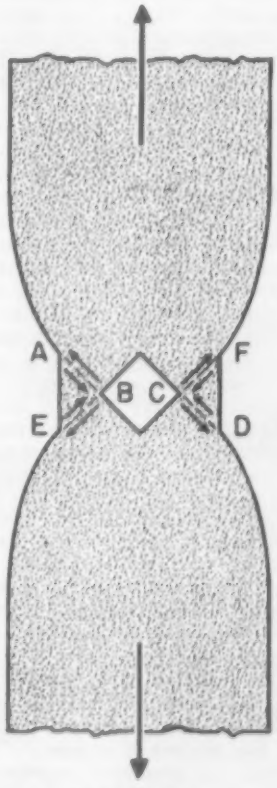

Figure 4

The conclusion is, then, that the Griffith energy principle can only be applied to fully or substantially brittle fractures; ductile fractures are quite outside its scope.

In arguing the applicability of the elastic energy release principle to ductile fractures, occasionally the point has been made that, if the specimen is long enough, the elastic energy stored in it should be sufficient to produce rapid crack propagation even if the energy absorption of the crack is as high as it is in typically ductile fractures. The answer to this is that a *fast* fracture is not necessarily a *brittle* fracture (i.e., a fracture involving very low energy absorption). Any ductile fracture can be made to run fast, at least from a certain stage onwards, if the specimen is connected in series with a large spring (or, what is the same, if the specimen is long enough). It can be shown that the condition for a ductile fracture to become a fast fracture is not eq 2 but equality of the *second* derivatives of W and U.*

The Writer's Crack Propagation Condition for Brittle Fracture in Normally Ductile Steels

As mentioned in the first Section, the present writer has suggested that brittle fracture in ductile steels may obey the crack propagation condition

$$\sigma \approx \sqrt{Ep/c} \qquad (1)$$

which results if, in the Griffith eq 10, the surface energy α is replaced by the surface plastic work p. It can be obtained by starting from the Griffith principle of elastic energy release eq 2 and equating the free energy required for producing unit area of the crack wall to p instead of α.

The first question is: Can the Griffith energy principle be applied to a fracture process that involves plastic deformation? It was seen in Section 2 that the Griffith equation can be derived from the elastic stress concentration factor of the crack; however, can this be done if plastic deformation takes place and redistributes stresses at the tip of the crack? The Neuber principle, mentioned in Section 2, shows that the

* To be published elsewhere.

stress concentration factor can be calculated approximately on the basis of the classical theory of elasticity if the plastically deformed region is small compared with the length of the crack. In this case it can be treated in the manner explained in connection with Fig. 2: the stress concentration factor is the same as that of a crack in a purely elastic body with a tip radius equal to half of the diameter of the plastically deformed region. In fact, this case is basically the same as that of the completely brittle material in which, in order to take into account the atomic structure of matter, the Neuber theorem had to be applied to the region at the tip of the crack in which the stress distribution flattens out owing to the presence of the maximum in the force-displacement curve Fig. 1. The only difference is that in the Griffith case the diameter of the non-Hookean region is of the order of the interatomic spacings, while in the brittle fracture of steel it is about twice the thickness t of the plastically deformed layer at the surface of the crack. According to the Inglis eq 13, the stress concentration factor is then

$$\eta = 2\sqrt{c/t} \qquad (17)$$

X-ray measurements indicate[1] that t is of the order of 0.2 to 0.4 mm in low-carbon steels broken not too far above or below room temperature.

In the Griffith theory, the tensile strength of the specimen was obtained by dividing the molecular cohesion by the stress concentration factor. What is the quantity corresponding to the molecular cohesion in the brittle fracture of steels? The clue is given by the important observation[7] that in steels the crack does not propagate continuously: before it has broken through a grain boundary, unconnected small cracks arise in grains ahead of the tip of the main crack. This shows that the brittle strength of steel cannot have the order of magnitude of the theoretical strength (molecular cohesion); in fact, it must be quite low if independent fracture processes can start ahead of the main crack at points where the stress cannot be much above the yield stress. This may be due to the presence of numerous invisible cracks scattered in the material; or to the well-known fact that plastic deformation can produce high microscopic internal stresses and subsequent crack formation. It seems that the cleavage strength of the material at the tip of the crack is not, or not much, higher than the ordinary brittle strength of steel obtained experimentally as the stress at which brittle fracture occurs. Since the cleavage strength of steel depends on the plastic strain which is difficult to estimate in the small region around the tip of a crack, only a rough idea at its magnitude can be obtained; it is probably somewhere between 100,000 and 200,000 psi for a low-carbon steel. For an applied tensile stress σ of, say, 20,000 psi, therefore, a stress concentration factor between 5 and 10 would be needed in order that the cleavage strength σ_m may be reached at the tip of the crack. If the thickness t of the cold worked layer in eq 16 is taken as $^1/_{100}$ in., the necessary crack length, given by eq 17,

$$c = \frac{1}{4}\,t\cdot\eta^2 = \frac{1}{4}\,t\left(\frac{\sigma_m}{\sigma}\right)^2 \qquad (18)$$

is between $^1/_{16}$ and $^1/_4$ in.; for a tensile stress of 10,000 psi the stress concentration factor is four times higher, and the necessary crack length is between $^1/_4$ and 1 in. These orders of magnitude appear quite reasonable in the light of experimental observations: brittle fracture does not occur around room temperature unless a notch or crack of this order of magnitude is present.

The last question to be discussed is whether eq 1 represents a sufficient as well as a necessary condition of crack propagation. At this point a significant difference appears between the fracture, say, of glass and of low-carbon steel. The rise of stress at the tip of a crack in glass is not limited by plastic deformation; in steel, however, the stress at the crack tip cannot exceed the yield stress multiplied by a plastic constraint factor which probably has a value between 2 and 3.[1] If, therefore, the cleavage strength is higher than two or three times the yield stress Y in tension, the tensile stress at the tip of the crack cannot reach the fracture level, no matter how high the stress concentration factor of the given crack in a purely elastic material would be. An additional point of great importance is that the yield stress of steel increases with the rate of deformation more rapidly than the yield stresses of most metals; between the usual rates of "static" tests and the fastest rates at which measurements could be carried out it seems to increase by a factor approaching 3. It seems that, in typical cases of brittle fracture in low-carbon steels, the velocity increase of the yield stress is the salient feature of the phenomenon of brittle fracture. Although cleavage fracture can arise at slow deformation rates, it then requires so much plastic deformation for producing the necessary plastic constraint and strain hardening[8] that the resulting cleavage fracture is anything but brittle; its energy absorption may be almost equal to that of a ductile fracture. Typical brittle fracture in a low-carbon steel, therefore, can occur usually only after the crack propagation has reached a sufficiently high velocity; in laboratory experiments, the fracture is almost always initiated by some ductile (fibrous) cracking, accompanied by considerable local plastic deformation.

It can be said, therefore, that a characteristic feature of brittle fracture in ductile steels is the enormous decrease of the crack propagation work with increasing velocity of the crack. The crack propagation condition eq 1 may well be fulfilled for a rapidly running crack with its low value of p but not for a stationary crack, the propagation of which may require, per unit of crack length, an energy of a higher order of magnitude. In such cases, cleavage fracture must be initiated in laboratory experiments by large deformations producing strong plastic constraint and strain hardening, and usually some fibrous cracking; the plastic deformation may have to extend across the entire specimen, so that the yield load has to be reached before cleavage cracking can start. After a cleavage crack has arisen, it may accelerate rapidly provided that the condition eq 1 is satisfied, so that there is sufficient elastic energy released during the crack propagation to increase the kinetic energy around the running crack. Consequently, it may be assumed that eq 1 represents a necessary and sufficient condition for the fast and, therefore, brittle propagation of a cleavage crack; in order to initiate a cleavage crack and accelerate it by static loading, however, an additional condition may have to be satisfied, demanding a certain amount of plastic deformation for producing plastic constraint and strain hardening.

It should be remarked that many service fractures seem to start without significant plastic deformation in spite of static loading. An interesting suggestion for explaining this has been put forward by Wells;[8] another possibility will be discussed in a subsequent paper.

This paper represents an expanded version of remarks that were stimulated by the work done under Office of Naval Research Contract No. N5ori-07870, and contributed to the Conference on Brittle Fracture Mechanics held at the Massachusetts Institute of Technology on October 15 and 16, 1953, under the auspices of the Committee on Ship Structural Design, advisory to the Ship Structure Committee, National Research Council.

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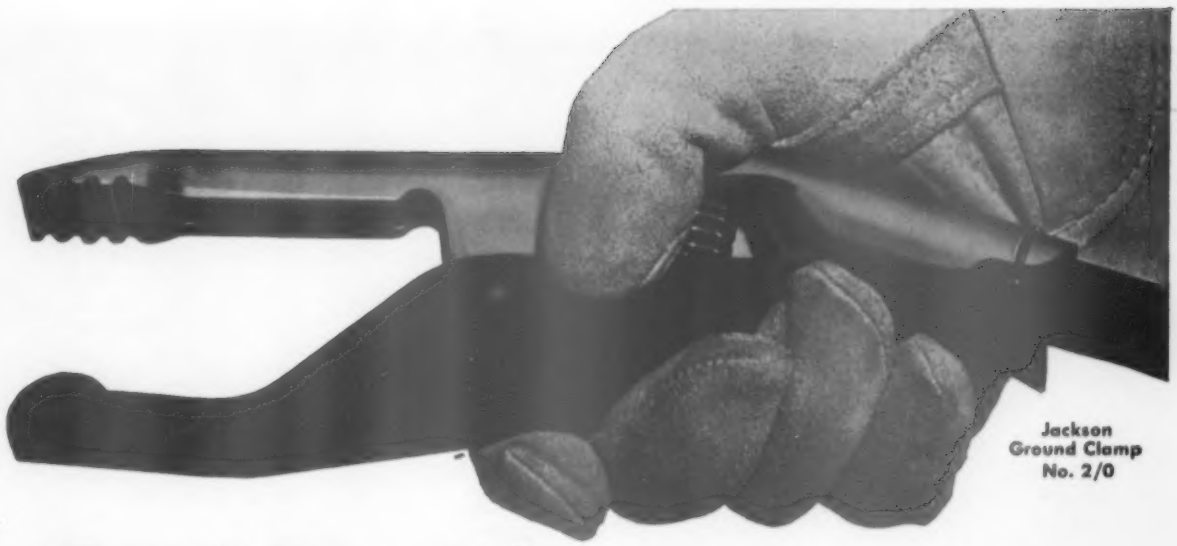

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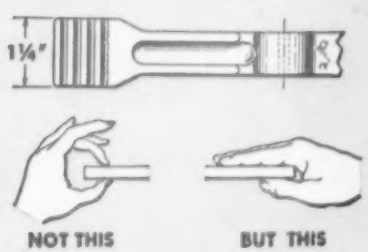

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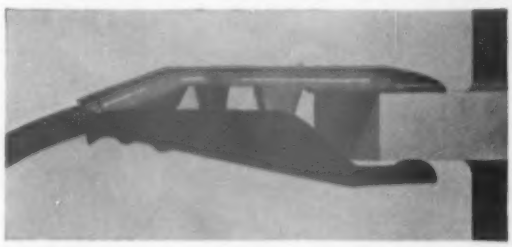

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